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Engineering Practice to Avoid Stress Corrosion Cracking

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ADVISORY GROUP FOR AEROSPACE RESEARCH AND DEVELOPMENT

(ORGANISATION DU TRAITE DE L'ATLANTIQUE NORD)

SYMPOSIUM ON THE ENGINEERING PRACTICE TO AVOID STRESS CORROSION CRACKING

Papers presented at the Structures and Materials Panel 29th Meeting
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THEME

Following a survey of the problems of aircraft construction relating to stress corrosion cracking the Structures and Materials Panel of AGARD decided that it would be useful to hold a symposium upon practical aspects of the problems as they can be controlled by design and production disciplines.

All the papers are directed to the practical aspects of the prevention of stress corrosion cracking, not only from a technical view but as a management function.

The symposium should therefore be of most value to senior design and production engineers and provide a forum for discussion at an entirely practical level.

As quality assurance and inspection departments have an active part to play between design and production engineers it is believed that senior engineers related to these activities should find the symposium of value and that they would be able substantially to contribute to the discussions.

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THE EXAMINATION OF SERVICE FAILURES

by

P. J. E. Forsyth A. I. M.

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Summary

The purpose of the first part of this paper is to provide a general introduction to the phenomenon of stress corrosion, briefly describing the conditions under which it occurs, and putting the aircraft problem into perspective in relation to fatigue.

The second part provides a practical introduction to fractures analysis, describing those features of stress corrosion cracking upon which recognition is based, and relating them to the microstructures of the alloys concerned. The observational techniques that are available are reviewed, and their application in various service failure investigations discussed.

Introductory lecture : The examination of service failures

Introduction to stress corrosion

The practical significance of stress corrosion, as with fatigue, is that cracks are formed that may seriously reduce the strength of load carrying members or components. Both of these forms of failure have occurred in a variety of aircraft materials over a number of years, and in considering the importance of the stress corrosion problem it is necessary to put it into perspective in relation to fatigue. In terms of numbers of failures of different items (which assumes that all have equal significance) our own records show approximately three fatigue failures to every one stress corrosion failure. However, significance cannot be simply measured in this way. A stress corrosion crack (for various reasons that will be described later) is less predictable in its growth rate than a fatigue crack, and the engineer who, on good advice, is prepared to keep a fatigue crack under surveillance, is far less inclined to fly an aircraft with a stress corrosion crack present in a sensitive position.

At this stage it is necessary to define the term stress corrosion, and to describe the conditions necessary for its occurrence. Stress corrosion cracking results from the conjoint action of a steady tensile stress and a corrodent, and cracks must necessarily start from exposed surfaces. The phenomenon has been observed in many of the commonly used structural materials, often with commonplace corrodents, and it is general experience that alloys based on most metals will crack if sufficiently stressed and exposed to the right environment. However, certain conditions of heat treatment may produce particular sensitivity in some alloys, and this will be discussed in more detail later. The fact that stress corrosion is not considerably more troublesome is because most of these specific environments are rarely met with in practice. However, it is an unfortunate fact that the strong aluminium base alloys are, in the main, susceptible to chlorides which are present in most environments. There is no complete explanation as to why certain compounds induce this form of cracking, while others, even those that may actively corrode the alloy in question, do not.

An alloy's susceptibility to stress corrosion cracking is measured in the laboratory in terms of stress, and time to failure under specified environmental conditions. This information is presented in the form of a life curve for various applied stresses, as shown in Fig.1. As with fatigue there is a strong stress dependence and it can be seen that the curve becomes assymtotic with respect to time, and a threshold stress may be determined that is analogous to a fatigue limit. In view of the length of time required to establish this threshold, it has been customary to test at high stress, often with accelerating corrodents, and if some arbitrary life is exceeded the material is pronounced to be safe for use. It is clearly more satisfactory to know the shape of the curve because the threshold stress is not necessarily related to high stress level behaviour, as can be seen in Fig.2.

In practice, all stresses will be raised by the geometric 'notches' present in the design, and the steady stresses that may be present fall into three groups:

(1) The working stresses imposed by the design

(2) The assembly stresses

(3) The internal or residual stresses in the metal itself.

These stresses will be additive with respect to direction and sign, and the low threshold strength that may exist in some materials imposes a severe restriction on the working stress that may be employed. Stress corrosion cracks develop across the direction of the resultant tensile stress, and the ease with which a crack grows may, in a wrought material, depend on the direction of that tensile stress with relation to the flow pattern in the alloy. Thus, in some aluminium alloys for example, the threshold strength for stress corrosion measured by stressing in the flow direction, (the longitudinal direction) may be above the yield strength, whilst the stress corrosion threshold strength across the grain (short transverse direction) may be as low as 1/6th of the longitudinal value.

At this stage it is necessary to define both the macro and microstructural features that are commonly associated with stress corrosion.

Structural features in alloys that influence stress corrosion susceptibility

It must be emphasised that stress corrosion is essentially a cracking phenomenon, and will therefore be influenced by the general fracture characteristics of the material. However, there is an overriding electrochemical component that depends not only on the environment, but the local chemical composition and physical state of the metal.

From this it will be seen that we can divide the structural features into two broad groups:

(1) Those that will affect fracture characteristics

and (2) Those that will influence electrochemical reactions

Those features that qualify for both groups are likely to have considerable influence on the phenomenon.

In most structural materials that concern the aircraft engineer the path of a stress corrosion crack will be intergranular, ie along the grain boundaries of the alloy. Those grain boundaries lying normal to the maximum tensile stress direction are most likely to fail because crack formation and growth depends on the local stress intensity as well as electrochemical dissolution. Having said this it will be seen that grain size and shape will influence the growth of cracks. The presence of extensive flat areas of grain boundary will provide easy crack paths, and any deviation that the crack tip has to make to follow the boundary path will reduce its stress intensity and therefore the growth rate.

We can now consider the typical grain boundary networks that we expect to find in commercial wrought materials. Some have fully recrystallized grains, in others only partial recrystallization occurs with the specification heat treatment, and there are other materials that are used in a cold worked condition. The fully and partially recrystallized structures that can occur are shown in Figs. 3, 4 and 5. These examples are all aluminium alloys, but the polygonal network of boundaries shown in Fig.3 is similar to that commonly obtained in a number of other materials including most steels.

Such a polygonal network of equiaxed grains will have no preferred direction for stress corrosion cracking, whereas the boundary arrangement shown in Fig.4, and to a much greater extent that in Fig.5, will provide practically continuous paths for crack growth. Thus, if the material is stressed across these paths crack growth is much easier than if it is stressed in a direction lying in this plane of weakness. This is always true for stress corrosion cracking and may also be so for tensile cracking if the grain boundaries have a particular mechanical weakness. It is necessary, at this stage, to remember the third dimension and consider what the solid grain shape is likely to be. In most working procedures the material will be extended mainly in one particular direction and be compressed at right angles to this direction. These are the two directions represented by the edges of the photographs. The material will spread in the third direction so that we can expect some less marked alignment of the planes along this direction. The three dimensional aspect ratio of the grains will depend on the methods of manufacture, but it is commonly found that the stress corrosion strength of the material when stressed in this third or transverse direction lies somewhere between the strength for longitudinal and short transverse.

In addition to the grain boundary networks the next important feature is the stringers or elongated groups of intermetallic particles that are part of the constitution of many materials. Such stringers can be seen in Figs.3, 4 and 5 and these features are known to reduce fracture toughness in the short transverse direction. These stringers are oriented in the flow direction of wrought materials and often lie along grain boundaries so that they may encourage the extension of a stress corrosion crack if the stress intensity is high enough.

Other stringers and inclusions of foreign matter are sometimes found in materials and if these should lie in a sensitive position with regards to the design they may encourage an early failure.

We have, up to this stage, discussed only relatively coarse microstructural features that could be seen clearly with the optical microscope, but which have a strong influence on the stress corrosion behaviour through the fracture characteristics of the material. It has been stated that in most structural alloys stress corrosion cracks follow intergranular paths, and it is now necessary to consider these grain boundaries in more detail. It is on this microscopic scale that the electrochemical dissolution occurs and it is generally agreed that for reasons of local composition variations anodic attack may occur at, and be confined to, the grain boundary itself. In some cases a distribution of precipitate particles of a secondary phase may provide the anode which will be necessarily small in relation to the large cathode area of the grains themselves and the mouth of the crack. It is not proposed to deal in any detail with the possible mechanisms of failure and it is sufficient to say that the dissolution of grain boundary material coupled with the necessary stress intensity to fracture any 'bridges' between the points of dissolution may be cited as a plausible mechanism for crack extension. However, there are ideas and theories which incorporate into the model the hydrogen that will form near the crack tip, and could be adsorped at the tip, as an embrittleing agent.

The different stress corrosion sensitivities that can be induced in many alloys both aluminium and steels by heat treatment are generally the result of variations in the fine precipitate structure, and not major rearrangements of the coarser features that have been discussed.

Having described the phenomenon of stress corrosion, and the features of the materials involved, it is now necessary to examine some service cases, giving the diagnostic features as observed on the fracture and in micro-sections.

Examination of service stress corrosion failures

Techniques

The number of techniques available for the examination of service fractures has increased over the last few years, and this, coupled together with a better understanding of the various forms of fracture, has made diagnosis more positive. The examination of the fracture is a

complementary exercise to the examination of the microsection and the two should be used together to obtain complete information regarding the relationship between fracture path and microstructure.

The examination of the fracture will naturally precede any sectioning, and it is in the field of fractography that considerable advances have recently been made.

Twenty years ago fractures were rarely examined at a higher magnification than a hand lens or binocular microscope would provide perhaps up to 100 diameters, and a diagnosis was made on the basis of this superficial examination and perhaps, a microsection. The higher powers of the optical microscope had been used for many years on basic studies of cleavage of single crystals at a magnification up to approximately x 1000, but nobody had thought it worthwhile to examine the more irregular surface features of service fractures in commercial alloys. Since then such observations have been made and the science has developed and has been applied to fatigue failures, then stress corrosion, creep and ductile and brittle tensile fracture. With the advent of the electron microscope and the development of replication techniques, notably evaporated carbon film replicas, the magnification that one can employ has increased by a factor of 20 coupled with a very much greater depth of focus. More recently the scanning electron microscope known as the 'Stereoscan' has become available, and with this instrument the fracture can be directly viewed without any replication problems. As with the conventional electron microscope great depth of focus is obtained and the useful magnification is about the same, ie ~ 20,000 x. With the conventional electron microscope using carbon replicas, the magnification limitation is imposed by the granular structure of the carbon and not by deficiencies in the microscope. With the scanning microscope using the actual specimen the limitations on magnification are, at present, within the instrument itself.

As with any other application of microscopy, an object or surface should be examined first at low magnification and then with increasing magnification as further resolution of detail is required. The simpler optical methods are without doubt still the most useful, and it is true to say that in most instances a general diagnosis can be made without further aids.

The more advanced methods may be required under the following circumstances:

(1) Where a fracture surface is badly damaged and only very small areas, perhaps lying in crevices, are available for examination.

(2) For thin sheet or other small fractures such as wires.

(3) Fractures where the suspect origin is very small.

(4) For determining the presence of fine detail such as the finer striations on some fatigue fractures.

(5) Where large depth of focus is an advantage.

The various features on stress corrosion fracture surfaces will be described by reference to service failures where the relevant conditions leading to failure will also be discussed, but before considering these in any detail a general comparison of stress corrosion fracture and fatigue fracture will be made. These are the two forms of service failure where discrimination is required, and very often the first question that the metallurgist is asked is, "are you sure that this failure is stress corrosion and not fatigue"? There is some justification for this initial doubt because in several respects they are similar, as shown in Figs.7 and 8.

We find that:

(1) There is very little general plastic deformation in either

and (2) The growth crescents are similar.

From this point we must progress to a more detailed study of the characteristics of stress corrosion fracture.

As Figs.7 and 8 show both the areas of stress corrosion or fatigue are generally smoother in texture than the rest of the fracture. The area of tensile rupture is in both cases rougher because it is more disturbed by plastic deformation than either the stress corrosion or fatigue areas. The occasional banding which appears with both stress corrosion and fatigue results from changes in growth rate, and in both of the cases shown bursts of faster fracture are followed by more settled periods of crack growth. Figs. 9 and 10 show typical stress corrosion and fatigue fracture surfaces at a few magnifications, as they would appear through a hand lens where it can be seen that although they are both more highly reflecting than the tensile rupture area, there are considerable differences in surface features. The stress corrosion crack has been developing along the grain boundaries, sometimes along parallel boundaries (consider Fig.5), and exfoliation of the 'plate like' grains occurs. On the other hand, the fatigue crack has developed along crystallographic planes to produce this typical facetted structure. Thus in an alloy where stress corrosion and fatigue follow these different paths their diagnosis is simple. Where they follow the same paths a more detailed examination will be required for other features that will be described later. The exfoliation of grains is a characteristic of wrought material where the

grains are 'plate like' and splitting, as illustrated in the microsection shown in Fig.11, has been occurring. At higher magnifications these relatively flat areas may be resolved into the component 'subgrains' where the boundary orientation differences are sufficient to 'pull in' the main boundaries. The contours of these exposed subgrain boundaries are shown in Fig.12.

If the main grains are equiaxed no general planes of weakness are observed, and a stress corrosion fracture in such a material will have an equiaxed granular appearance, as will be seen in some of the stress corrosion fractures in steels.

Some service stress corrosion fractures that illustrate the features that are of diagnostic value

The first case, which will serve to illustrate the main characteristics of stress corrosion fracture is a centre section forging, an integral part of the structure of an aircraft, a frame with overall dimensions approximately 8ft x 3ft x 6in thick, part of which is shown in Fig.13. This forging was manufactured in the high strength aluminium-zinc-magnesium-copper alloy to U.K. specification DTD 683. The short transverse direction was perpendicular to the main plane of the frame, and the working stresses in this direction were considered to be negligible. However, stress corrosion cracks have developed as a result of the residual stresses in the component, occurring as illustrated in Fig.13, arrow A. Fig.14 shows such a crack after it had been opened for examination. The typical features of a stress corrosion fracture are present, the relatively smooth 'thumbnail' origin and the occasional tide marks that we have already discussed. At a higher magnification, the exfoliation could clearly be seen. In some stress corrosion fractures, depending on the environmental conditions that existed in service, we find that the area near the origin may be stained or even corroded to an extent that has produced discernable corrosion product. This has happened in this case. Conversely, the further the crack grows towards final failure the less time for corrosion to occur on the fracture surface, and the brighter it appears. This is not an invariable rule because in some circumstances it appears that cracks dry out at their mouths and water may be held by capillary attraction at the crack root where the crack opening displacement is smaller. This results in the occasional bands of staining that are found well away from the origin of the fracture; a feature that can also be seen on some corrosion fatigue fractures. After the fractographic examination this fracture was sectioned and Figs.16 and 17 clearly show the intercrystalline nature of crack growth. They also indicate how exfoliation of the grains comes about, where the parallel cracks can be observed as mentioned earlier. Stress corrosion in aluminium alloys is commonly accompanied by intracrystalline corrosion at the mouth of the main crack where considerable material may be removed and a pit formed. Although it is commonly found in areas of active corrosion, particularly under contact corrosion conditions, it is not a pre-requisite for stress corrosion, but can occur during the growth period. Its character is of considerable interest†, see Fig.17, where the corrosion stops short of the grain boundaries, often leaving them outlined and wholly unattacked.

As regards the failure itself, there are several points to consider:

(1) The source of the stress

The high residual stresses resulted from the large bulk of material that had been quenched and the considerable amount of subsequent machining done to form the fork fitting, etc. These failures are a legacy of earlier days when firstly cold, and later warm water quenching was used. The present specification for this type of material calls for a boiling water quench – in other circumstances where step quenching in molten salt at 180°C is feasible, this might be even more preferable and the residual stresses reduced even further.

(2) Design in relation to environment

The shape of the centre section was such as to trap condensate water at a position where high residual stresses existed. The presence of steel parts in the near vicinity may have aggravated the situation by rusting and contaminating the condensate.

(3) Other significant points

The protective scheme of anodizing followed by painting was very good, but eventual breakdown had occurred at several points. Although it could not be proved, it is likely that this happened at points where oxide inclusions were exposed at the machined surface, and where the subsequently formed anodic film might be defective. A certain number of these inclusions have been detected ultrasonically and subsequently exposed and examined. Fig.18 shows an inclusion exposed on a fracture surface and Fig.19 a similar type of inclusion in a microsection. Such defects are less likely to occur in more modern material, and present day inspection techniques would reveal their presence.

When a stress corrosion crack has grown until the stress intensity at its tip exceeds the critical value for f st fracture, then complete failure will ensue. In some practical cases where the main operative stress is a residual one the growth of th c ack may re ax the loa and t e stress intensity may never reach the critical value for fast fracture. This behaviour has been observed in several cases.

Other aluminium alloy failures

Another service stress corrosion failure, an undercarriage leg forging, again in the aluminium-zinc-magnesium-copper alloy, DTD 683 is shown in Fig.20. A longitudinal crack about 14" long was discovered in this leg. When the crack was opened stress corrosion origins were found at the dowel hole as illustrated in Fig.21. Again, the typical 'thumbnail' area can be seen followed by the fast fracture area. In this case there was some corrosion resulting from the contact of a bronze sleeve within the bore of the leg. This sleeve was pegged by a steel dowel that screwed into the hole, the thread of which is visible in Fig.21. The material at the origin was therefore in contact with the bronze and in the near vicinity of the end of the dowel. The sleeve had been cadmium plated but the coating had been partly scuffed off during assembly. The clearance between the end of the dowel and the hole formed a crevice where moisture had been trapped and corrosion occurred. The stresses were thought to be a combination of working and residual. Although the Al-Cu-Mg alloys are generally considered to be less prone to stress corrosion in comparison with the Al-Zn-Mg-Cu alloys, a number of failures have been experienced, particularly under active corrosion conditions or conditions aggravated by contact corrosion. The following two cases may be of some interest, both were hydraulic cylinders where corrosion was apparently aggravated at clamping rings. Fig.22 shows the outer surface of one of these cylinders where the general state of corrosion can be seen with the presence of corrosion pits in a line corresponding to the edge of a clamping ring. Fig.23 shows the exposed fracture revealing the fibrous texture of the extrusion from which the part was manufactured, and Fig.24 confirms the intergranular nature of the cracks and the presence of intragranular corrosion that formed the pit. This material was 2024-T4 where a recognised short transverse stress corrosion susceptibility occurs in the naturally aged condition.

The second case is somewhat similar, but with the complication of a combined stress corrosion and fatigue history. Again an aluminium-copper-magnesium alloy hydraulic component where some contact corrosion was found on the outside surface of the cylinder, as shown in Fig.25. Fig.26 shows an exposed crack where areas of different texture can be seen. Figs.27 and 28 show the nature of these differences as revealed at higher magnifications. The stress corrosion is revealed as a typical intergranular surface and the fatigue is confirmed by the presence of crack growth striations.

Other specific features that are associated with stress corrosion

Assembly stresses have already been mentioned, and there has been considerable experience of trouble from these stresses. The connection of stiff structural members through angle or 'pi' sections is a common cause of trouble, and in these situations high bend stresses are often experienced across the grain flow of the material. This may be aggravated by stress raisers such as rows of rivet or bolt holes.

Another common source of trouble is the interference fit bush which may be bronze or steel, and should be cadmium plated and assembled with a jointing compound. Where adverse tolerances occur high local stresses may result, and again this situation usually exploits the short transverse weakness of the material. Again, in a similar category, the use of taper bolts in reamed holes can, if the torque is not controlled, cause stresses above the stress corrosion threshold in the short transverse direction in the material.

These situations may be coupled with another feature, ie., the existence of heavily sheared surfaces by various forms of machining. Fig.29 shows such a surface structure produced by drilling and reaming. The residual stresses may be high enough to cause cracking in such a structure, and in fact the special sensitivity of such a feature to stress corrosion forms the basis of the so called 'cut edge' test that is used as a rapid and simple test for material susceptibility. If a hole, machined in a manner to produce this feature is loaded by an excessive interference fit, then cracking may well occur, although in some circumstances the stress intensity may reduce as the crack grows away from the hole.

The simple shear that can result in deformation of the exposed end grain can also be produced during reworking operations where dressing out of corrosion product is being attempted. In some cases 'persistent' slip band damage very similar to fretting damage has been produced by the use of rotary files. This form of damage is shown in Fig.30. This would most likely constitute a stress corrosion hazard although we have no unambiguous evidence of this happening. There is no doubt that these features are the result of heavy machining cuts or the use of blunt tools.

Stress corrosion in other structural materials

The diagnosis of stress corrosion in aluminium alloys is relatively straightforward because of its intergranular path. The only other likely form of fracture in commercial alloys that would follow the same path would be creep, and this form of failure could usually be determined from the working conditions.

When considering the other important structural materials we find that in most cases the problem of diagnosis is complicated by two factors:

(1) There are other forms of failure, in particular hydrogen embrittlement that may have similar fracture characteristics.

(2) Stress corrosion cracks and fatigue cracks may follow the same paths.

These points are best illustrated with service examples.

Figs.31 and 32 show the failure of a nickel-chromium-molybdenum high tensile steel aircraft wheel axle. The fracture exhibits growth markings and there is no doubt that it is some form of delayed fracture. The intergranular nature of the fracture suggests stress corrosion. In this particular case extensive rusting had occurred as a result of inadequate protection after reconditioning.

In many cases the corrosion is not so obvious and the question arises as to whether the failure is stress corrosion or possibly hydrogen embrittlement which can also follow an inter granular path. Opinions vary on the hydrogen content required to cause cracking in a particular steel, and in any case values that are considered marginal are often obtained by analysis. Under these circumstances diagnosis can be very difficult. Various workers have suggested that subtle differences in the fine detail on the grain boundary facets can be identified. It has been suggested that the presence of small pores at the grain boundaries, which are revealed as pits on the boundary facets, are a feature of hydrogen embrittlement. Other workers have stressed the importance of fine features that indicate the degree of mechanical, as opposed to corrosion, separation that has occurred. As stress corrosion itself is a combination of mechanical rupturing and corrosion separation it would seem to be impossible to satisfactorily separate the two forms of failure on this basis.

The fact that hydrogen cracks initiate under conditions of hydrostatic tension means that internal origins are likely, and if such are found this would eliminate stress corrosion as a possible cause of failure. A close study of the local directions of crack growth might establish this point. However, if sharp notches or even fatigue cracks are present it would seem possible to get the conditions for hydrogen crack initiation at the tip of the crack and under these circumstances it might be impossible to differentiate between stress corrosion and hydrogen embrittlement.

Fig.33 shows a fracture surface in a 18% nickel maraging steel casting. The small casting had been part machined exposing coarser core grains and had been cadmium plated and out baked. 0.125 p.p.m. of hydrogen was analysed in the near vicinity of the fracture, the component had only been loaded for a few hours and the cadmium plate had not been disturbed. The crack had occurred at a stress concentration and apart from the general intergranular nature of the fracture there were one or two cleavage facets in evidence. Fig.34 shows both intergranular and cleavage cracking in a microsection. From a general consideration of the circumstances of this failure it was classified as hydrogen embrittlement.

A further case where corrosion had been evident on the component surface, in this case a steel draw bolt in B.S.S.95, was studied in some detail. The fracture surface features are shown in Figs.35 and 36 where again the intergranular nature of the fracture path can be clearly seen.

Figs.37 and 38 show the brittle intergranular nature of a fracture in a small latch, the material being A.I.S.I.431 which approximates to the U.K. specification B.S.S.80. This steel may be embrittled by tempering in an incorrect temperature range with accompanying stress corrosion susceptibility, and the microstructure suggested that this could have happened. It was considered that stress corrosion was the most probable cause of the failure. It is interesting to compare the various photographs of intergranular fracture over the range of materials discussed.

Certain problems of diagnosis arise when stress corrosion, fatigue and even tensile fracture follow the same paths. Single phase copper base alloys can fail in an intergranular or transgranular manner depending on the environmental conditions, and it is fairly common to find both forms of cracking on one fracture. Because single phase alloys are generally ductile, overstressing can easily be detected. However, considerable problems arise when the possibility of corrosion fatigue exists. Under these circumstances the stress and chemical environment must be fully considered.

With two phase or duplex copper base alloys, or for that matter any alloys where the two primary phases are about equally divided, fracture analysis may be particularly difficult. In copper base materials the α phase may be more ductile and corrosion resistant whereas in titanium alloys where aluminium is present in the alloy the α phase may be both brittle and have a peculiar susceptibility to room temperature stress corrosion. This susceptibility can only really be revealed in precracked material where stress corrosion cracks can be made to grow if the stress intensity is high enough. It is not proposed to deal in any further detail with this important point as this will no doubt be dealt with in later lectures. However, it is necessary to consider the features of fracture in this type of material. If one phase is susceptible to a brittle form of stress corrosion cracking and the other is less susceptible or even completely insensitive then the crack can only continue to grow if there is some degree of mechanical tearing of the ductile phase, or if a fatigue stress is present. The higher the volume fraction of the susceptible phase the easier will the cracks grow and therefore the lower the stress intensity required to produce growth. The adjustment of volume fraction could be the result of varying composition so that the inherent susceptibility of the α phase might be altered as well. Considerable areas of cleavage fracture can be produced in such alloys by both stress corrosion or by fatigue conditions, and it is only in circumstances where the stress intensity is high enough that the fatigue striations are visible.

The stress corrosion paths found with some titanium alloys subjected to hot salt (600–800°F) is reported to be both intergranular and transgranular, although in these cases cracking can occur from plain surfaces, often from exposed grain boundaries or slip bands.

Fig.39 shows the texture differences on a fracture in titanium alloy 811 where the first part of the crack has grown by fatigue from a notch, the second stage is stress corrosion under 3% NaCl solution, and the last stage is fast fracture. At higher magnification the stress corrosion part of the fracture can be resolved into cleavage facets as indicated in Fig.40.

A recent service failure in Hylite 51 showed extensive cleavage areas together with connecting bridges that failed in a ductile manner as shown in Figs.41 and 42. In some areas there were also indications of rudimentary striations that indicated some form of intermittent crack growth. Fatigue striations produced in this material in the laboratory under known cyclic loading conditions were far more clearly and characteristically defined and this illustrates the difficulty of diagnosis that is experienced in many cases.

It can be demonstrated that slowly repeated loads applied to an aluminium alloy under the right environmental conditions can produce a striated stress corrosion fracture where the path is intergranular, at higher frequencies the path is transgranular and can be classified as corrosion fatigue. It is to be expected that if a stress corrosion crack has formed in a service component there will be a stage when the stress intensity at the crack tip resulting from the cyclic loads experienced in service will cause intermittent growth conditions. If both stress corrosion and corrosion fatigue cracks follow the same transgranular path then it may be extremely difficult to classify the failure. In this respect the basic understanding of the mechanisms involved, and the true contribution of the reversal of stress, is not yet fully understood and it is to be expected and hoped that advances will soon be made in this field that will aid the failures investigator.

Conclusion

The conditions that encourage stress corrosion and the diagnostic features that stress corrosion fractures show have been discussed. Aluminium alloys are, generally speaking, the easiest materials in which to diagnose either stress corrosion or fatigue. The steels are complicated by having both trans and intergranular stress corrosion cracking paths, and also by the fact that hydrogen embrittlement cracks may follow similar paths. Two phase alloys, both copper and titanium base, may be difficult, particularly where both stress corrosion and fatigue follow the same fracture paths.

The advanced observational techniques already described should be used wherever possible to substantiate the simple optical methods, but it should be emphasised that in many cases diagnosis is possible with only a hand lens, or at most a low power binocular microscope.

Acknowledgements

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Fig.41	Cleavage fracture facets with river markings and areas of ductile fracture in titanium alloy. Hylite 51. Electron micrograph x 1500
Fig.42	As for Fig.41 but scanning electron micrograph x 1500.

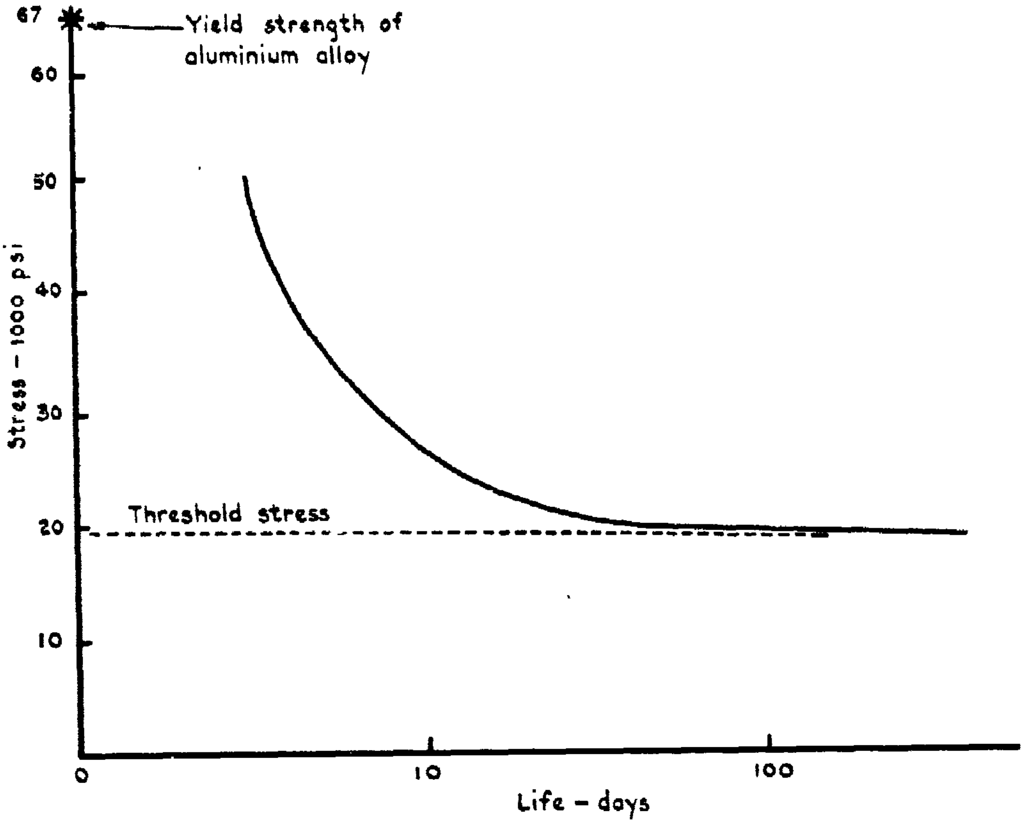


Fig. 1 Typical stress corrosion / life curve

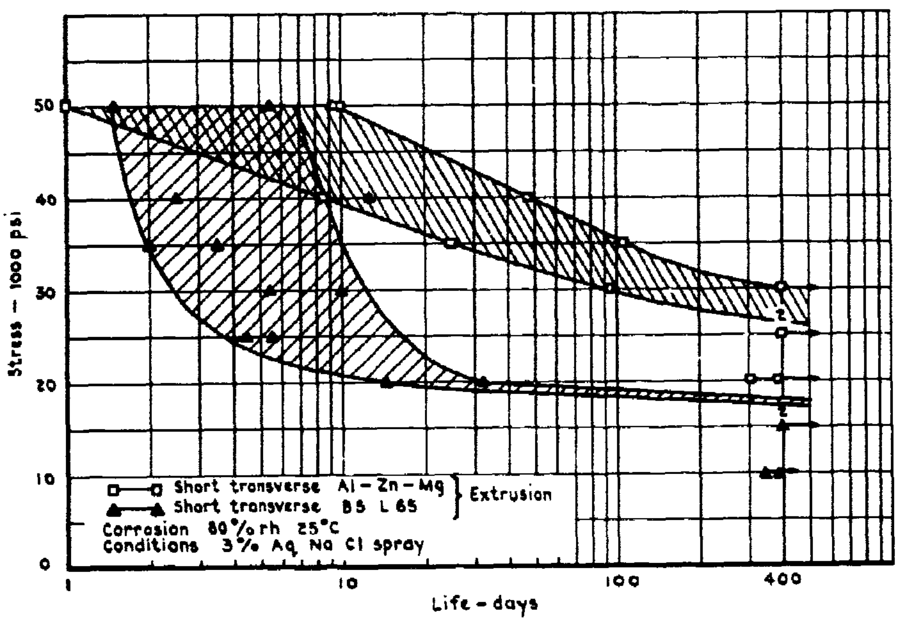


Fig. 2 Stress corrosion results of a boiling water quenched Al-Zn-Mg experimental alloy Compared with BS L65 alloy

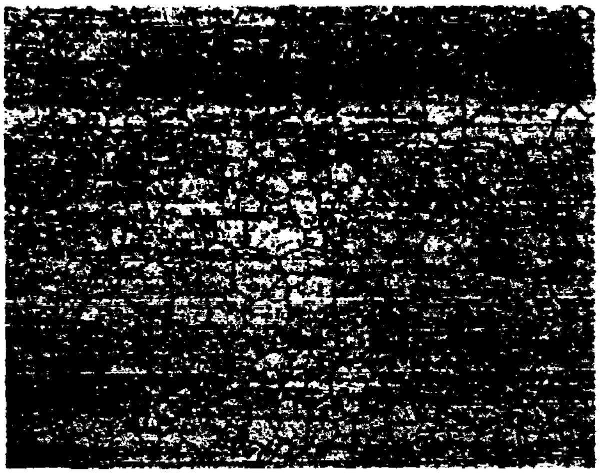


FIG. 3

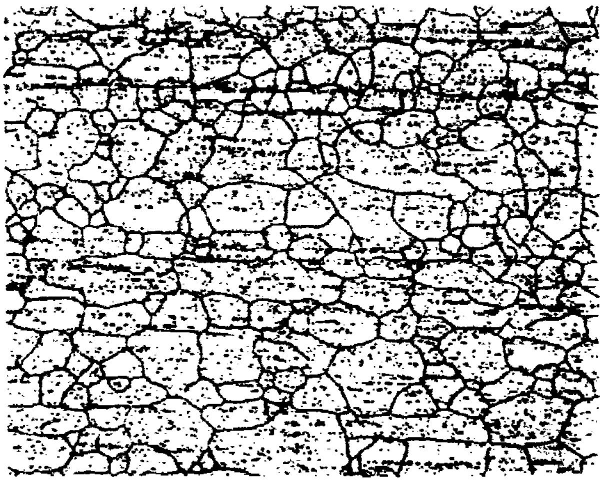


FIG. 4

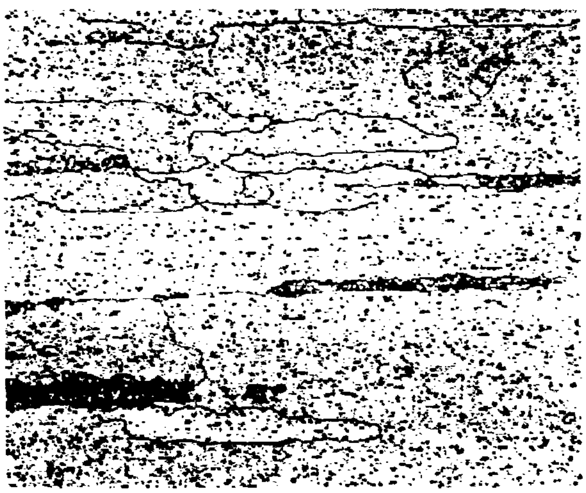
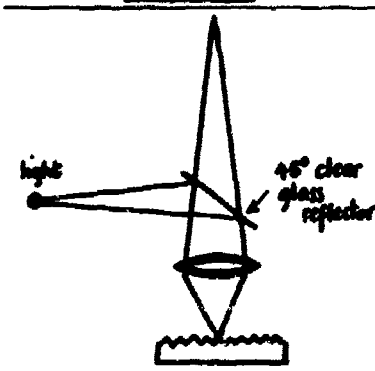


FIG. 5

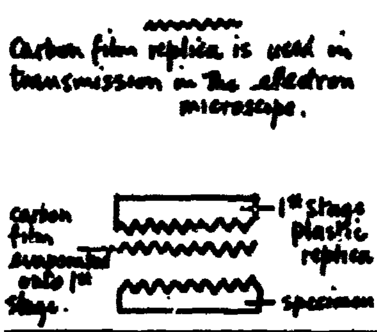
Fig.6 Fractographic techniques

Hand lens - oblique lighting, magnification usually about x 10. Large field of view, large illuminating source, eg. daylight reveals subtle differences in texture and colour.

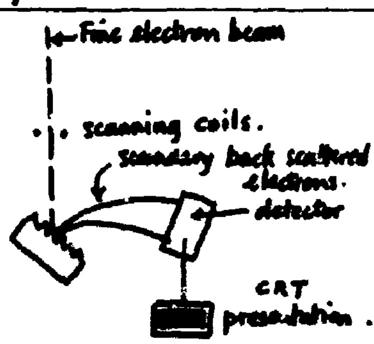
Low power binocular microscope, oblique lighting, magnification up to about x 100, better stereoscopic vision than hand lens, quick change magnification or zoom lenses available.



Conventional metallographic light microscope, vertical or dark field illumination, magnification up to about x 1000. Very limited working distance at the high magnifications and poor depth of focus. Better 'light grasp' with the higher numerical aperture objective lenses. This increases the range of angular tilt of the specimen surface that can usefully reflect back light to form an image. This apparatus will reveal most diagnostic features with colour differentiation of intermetallic particles etc.

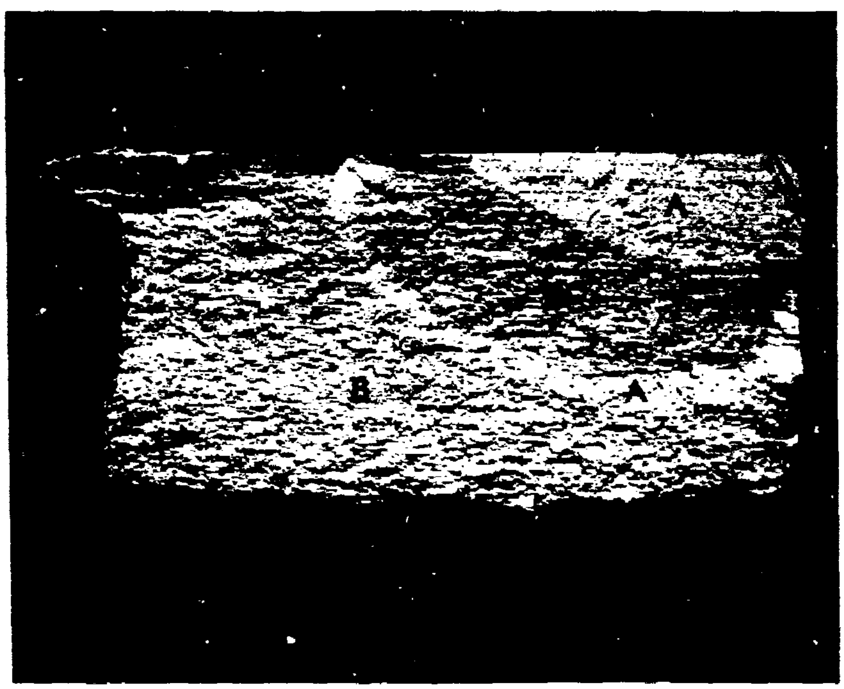


Electron microscope, magnifications up to the limit of resolution of ~ 3A but in practice limited by the internal structure of the replicating material, eg carbon - useful magnification limit x 20,000. Replicas must be used, except when the microscope is used in reflection, but this is rarely used for fractography. Considerable increase in resolution over the light microscope, increased depth of focus, enhanced contrast of topography by metallic shadowing. Possibility of artifacts in replication and no colour.



Scanning electron microscope or 'Stereoscan'. Direct use of the specimen, no replicas, useful magnification limit ~ x 20,000. Resolution and depth of focus comparable to that obtained with replicas in the conventional electron microscope. Magnification switching from macro to micro examination. Precise alignment of any spot on the fracture. Reveals secondary cracks in depth and exposed particles on fracture surfaces. Some shallow detail may be difficult to reveal compared with conventional electron microscope where the replicas can be shadowed.
Again, no colour differentiation.

FIG. 7



[illegible]

FIG. 9

FIG. 10

FIG. 11

FIG. 12

FIG. 13

FIG. 14

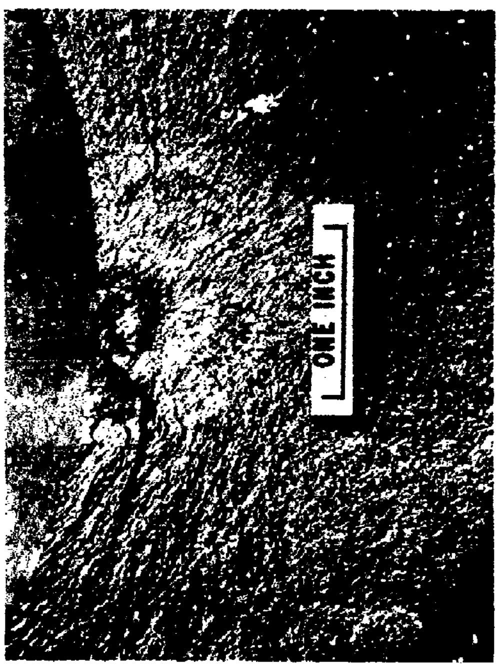


FIG. 15

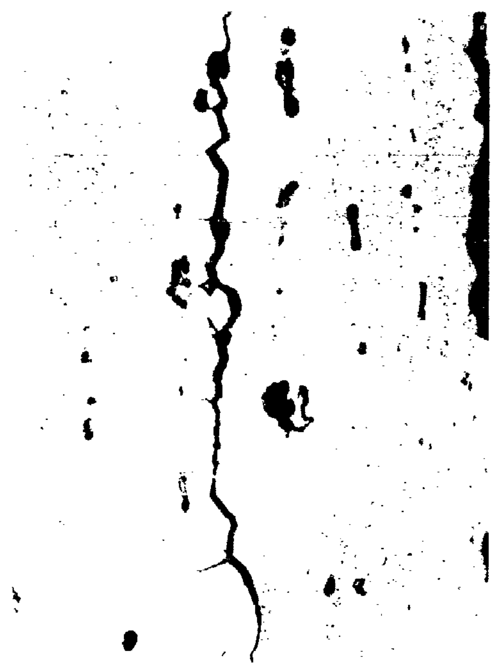


FIG. 16

FIG. 17

FIG. 18

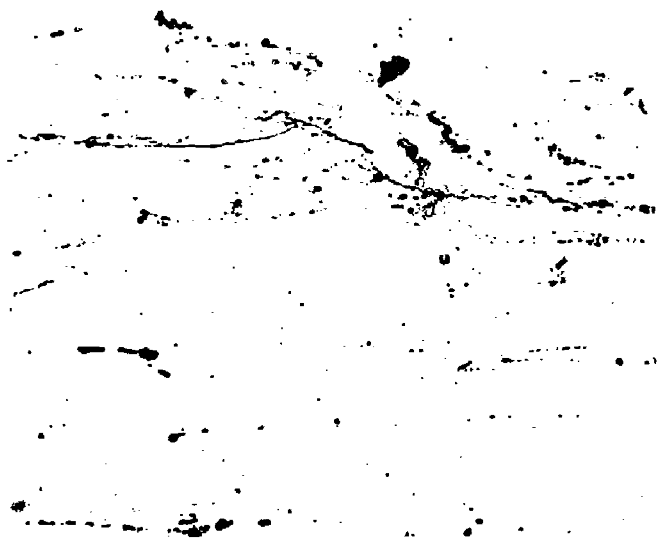


FIG. 19

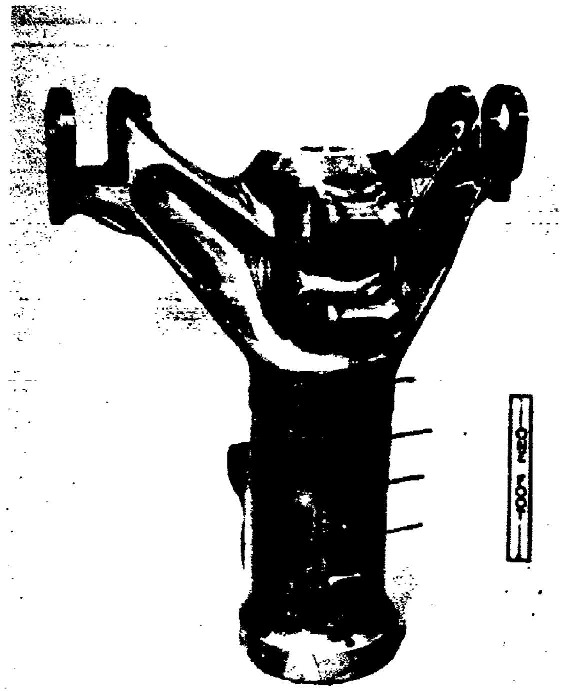


FIG. 20

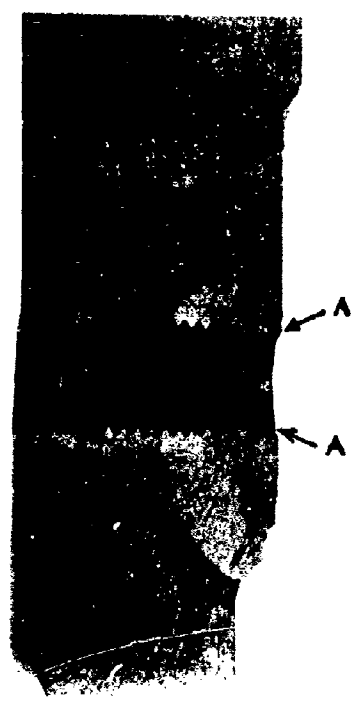


FIG. 21

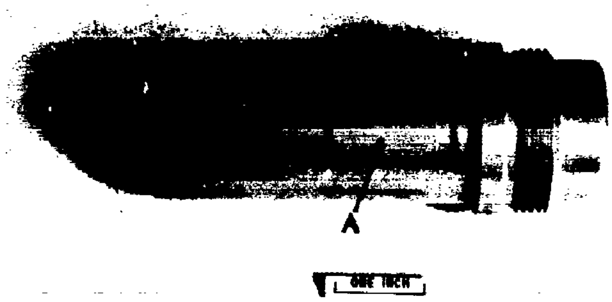


FIG. 22

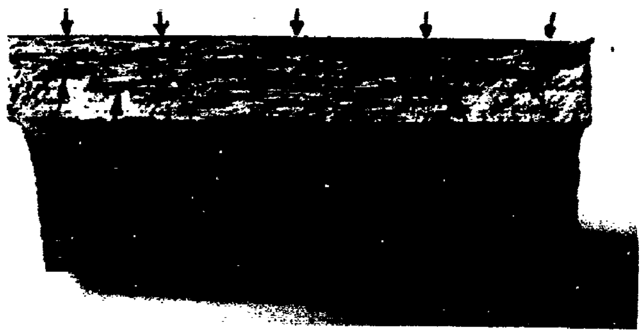


FIG. 23

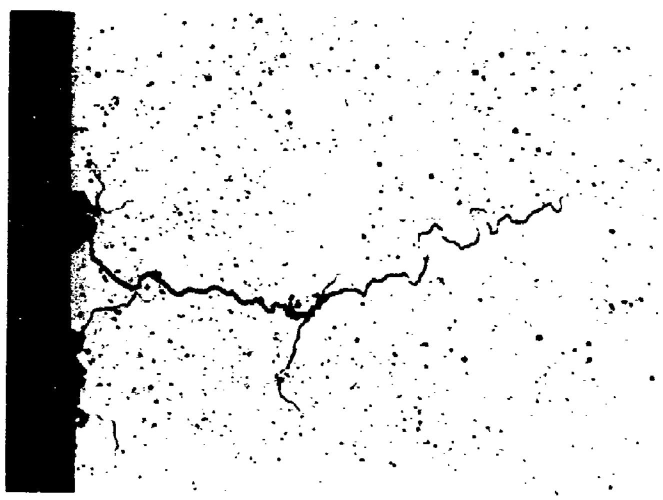


FIG. 24

FIG.25

FIG 26

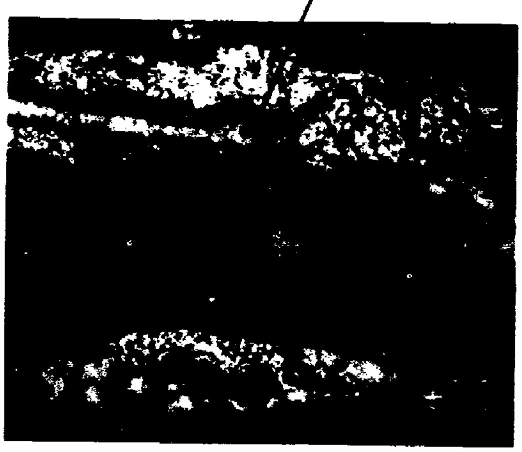


FIG.27

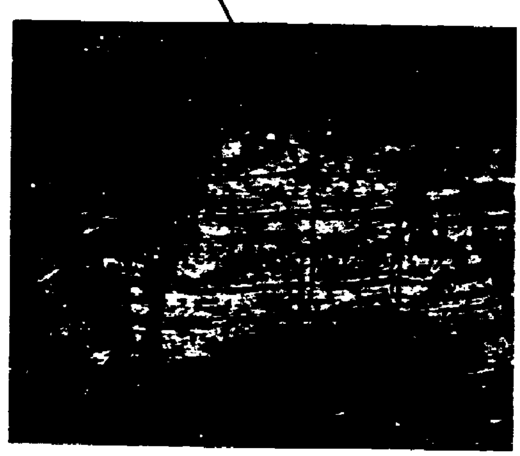


FIG.28

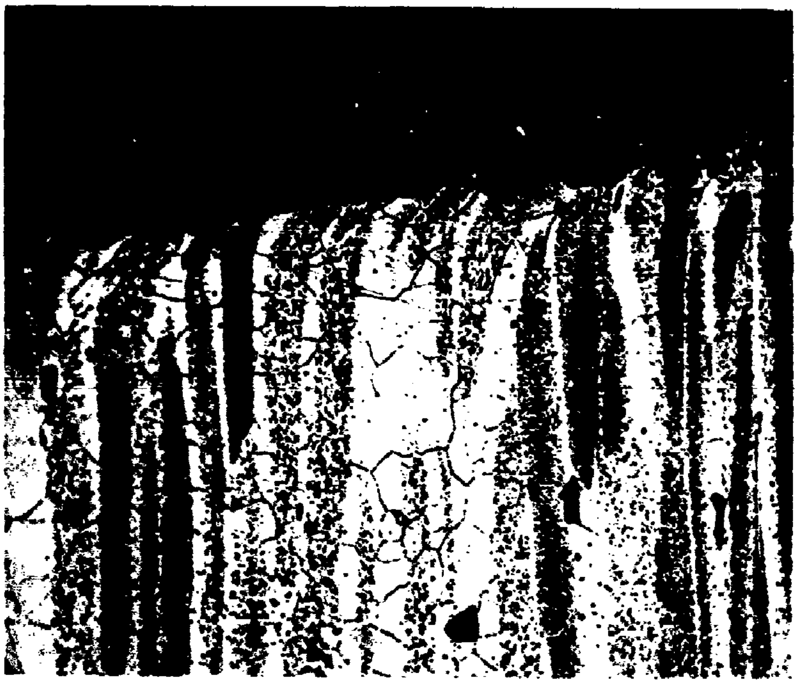


FIG. 29

FIG. 30

FIG. [illegible]

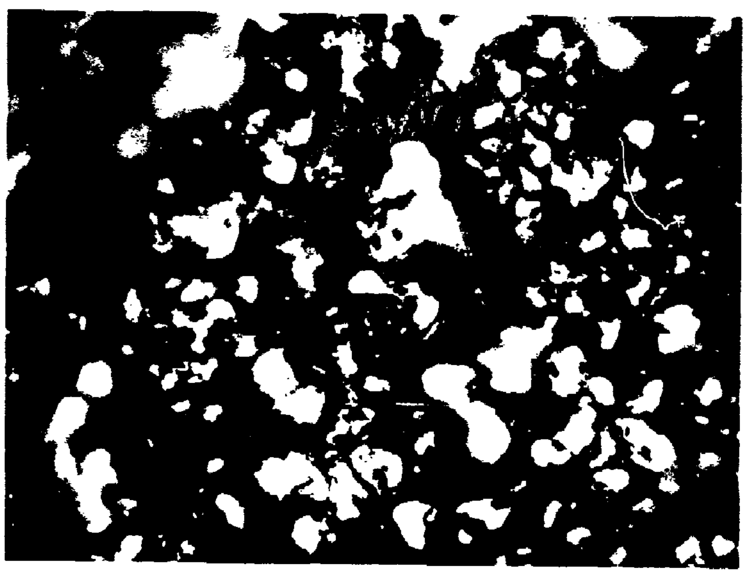


FIG. [illegible]

FIG. 33

FIG. 34

FIG. 35

FIG 36

FIG. 37

FIG. 38

FIG. 39

FIG. 40

FIG. 41

FIG. 42

THE RELATIVE INFLUENCE OF STRESS AND ENVIRONMENT ON TITANIUM ALLOYS

by

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Columbus, Ohio

THE RELATIVE INFLUENCE OF STRESS AND ENVIRONMENT ON TITANIUM ALLOYS

R. I. Jaffee

In common with all high-strength structural materials, titanium alloys are susceptible to stress corrosion cracking under particular circumstances, and stress corrosion cracking has been encountered in a number of the applications. Usually, the situation may be fixed by modification of the alloy, its condition, or the environment. In a few cases, the use of titanium is inadvisable.

The following sections of this paper will review the various sensitive environments for titanium and indicate the most feasible means, if any, of using titanium in the environment. The status of our understanding of the mechanism also will be commented on briefly.

SEA WATER

Titanium alloys are used for deep submergence vehicles and in various marine applications, during which they are exposed to sea water under stress. In many aircraft applications, particularly naval aircraft, exposure to sea water is anticipated, and resistance to stress corrosion cracking is necessary.

Tough titanium alloys are characterized by high values of fracture toughness, critical stress intensity factors K_{Ic} of 80-120 ksi$\sqrt{\text{in.}}$ Susceptible titanium alloys have K_{Ic} values as low as 20-30 ksi$\sqrt{\text{in.}}$ when loaded in salt water. Under steady load susceptible titanium alloys undergo slow crack growth over a period of hours forming flat cleavage-like cracks normal to the tensile stress until the intrinsic stress intensity value is reached (80-120 ksi$\sqrt{\text{in.}}$), and rapid, ductile shear fracture occurs over the remaining section. A plane strain condition such as is found in heavy sections, and a precracked condition, appear to be necessary to initiate the failure. Many titanium alloys which would be susceptible under plane strain are not susceptible under plane stress (e.g., thin section) conditions.

Control

The best means of dealing with the sea water stress corrosion problem for titanium alloys is to use non-susceptible alloys in a metallurgical condition in which susceptibility is minimized. In this way, very satisfactory combinations of strength and sea water immunity may be achieved.

The chief alloy contributors to stress corrosion cracking susceptibility are aluminum, oxygen, and hydrogen. Aluminum should be kept at 6% or below, oxygen at 1200 ppm or less, and hydrogen well below the 150 ppm maximum. Cracking generally occurs transgranularly through the alpha phase, and it has been found helpful to have beta phase distributed through the alpha matrix to act as crack stoppers. The presence of brittle transformation products in the beta phase, such as omega or eutectoid-formed compounds, are detrimental to stress corrosion cracking resistance. Isomorphous beta stabilizers, like molybdenum and vanadium, are most desirable. Examples of immune alloys are Ti-4Al-3Mo-1V and Ti-6Al-2Sn-4Zr-2Mo. The main production alloy, Ti-6Al-4V, is not susceptible, but marginally so, and must be used in favorable conditions to avoid susceptibility.

Metallurgical condition has an important influence on susceptibility. The main concern with alloys containing high aluminum is to avoid slow cooling and annealing through the 1000-1400°F region, where Ti_3Al forms as a coherent precipitate. Thus, mill-annealed Ti-6Al-4V is susceptible, whereas duplex-annealed or quenched Ti-6Al-4V is not. Care must be taken in stress relieving to avoid incurring susceptibility to sea water cracking. The fine grain size obtained through beta processing of alpha-rich alpha-beta alloys is favorable to resisting stress corrosion cracking. Such transformed beta-structures generally result in lower tensile ductility. The beta-processing treatment might be inadvisable if high tensile ductility is required, such as for sheet forming. The main idea is to avoid large alpha grain size. This would be particularly a matter of concern in the all-alpha alloys, like Ti-5Al-2.5Sn.

Mechanism

The mechanism of titanium alloy cracking in sea water is not understood, although a considerable amount of work has been done and is underway. It is generally agreed that the coplanar slip on $(10\bar{1}0)$ and $(10\bar{1}1)$ in the alpha phase promoted by aluminum over 6% is a contributing factor. However, the electrochemical reaction occurring at the base of the crack is unclear, and the roles played by chloride ion and hydrogen are in dispute.

N_2O_4

N_2O_4 is the oxidizer used in conjunction with hydrazine rocket fuels for many space and missile systems. Generally, N_2O_4 is contained in welded Ti-6Al-4V tanks. The condition where cracking was encountered was at

moderately elevated temperature, 105°F, and high wall stress approaching the yield strength of the tankage. The incidence of stress corrosion cracking in N_2O_4 was significantly increased when higher purity (red) N_2O_4 was substituted for lower purity (green) N_2O_4.

Control

The chief means of control of cracking in N_2O_4 was through con[illegible] of the environment. The critical difference between green and red N_2O_4 is a greater content of NO in t[illegible] [illegible]reen. Thus, N_2O_4 oxidizer now used has a specified content of NO of about 0.6% NO, and is designat[illegible] [illegible]ibited N_2O_4. It is also possible to achieve noncorroding N_2O_4 by addition of a small content of abo[illegible] H_2O, which effectively results in increased NO.

Another alleviating measure, which worked reasonably well with red N_2O_4, is shot peening the inside of titanium alloy tanks with glass beads. The residual compressive stress at the surface is very effective in reducing stress corrosion susceptibility.

Modifications in the alloy and its condition have not been effective in dealing with the red N_2O_4 stress corrosion problem.

Mechanism

N_2O_4 stress corrosion is a surface-initiated reaction, not requiring the presence of a notch or a precrack. It appears to involve the presence of HNO_3 or free oxygen when the water and NO content in the N_2O_4-H_2O equilibria approaches the anhydrous condition. Apparently the titanium protective oxide film cannot repair itself when H_2O is not present. The reasons for this are unknown.

METHANOL

Titanium alloy tankage for space and missile systems sometimes was pressure-tested while filled with anhydrous methanol, which has similar density and viscosity to the liquid fuels and oxidizers used. However, small amounts of halides or halogens, such as might be encountered as impurities or from contamination, render methanol an active stress corrodant for titanium.

Control

Modification of the methanol by addition of a small content of H_2O will render it innocuous. However, since methanol is not essential as a pressure-testing medium, the problem has disappeared by switching to another pressurizing liquid (probably H_2O).

Mechanism

As in the N_2O_4 case, precracks are not necessary to initiate failure. Methanol cracking appears to be a surface phenomena, probably involved with the reduction of interatomic bond strengths, due to the presence of adsorbed molecules. This requires fresh metal surfaces, such as occurring at the base of oxide film cracks. Thus, stresses sufficient to induce plastic deformation are necessary. In the absence of water, the reaction product between titanium and methanol is titanium methoxide, which is soluble, and not protective. The immunizing action of water additions makes it possible to repair the normal TiO_2 film once it is cracked.

HOT SALT

Jet engines ingest a small amount of sea salt when operating in marine environments. Sampling tests indicate it is possible to have salt deposits of about 0.1 mg/in^2 on titanium compressors during transoceanic services and as high as 1 mg/in^2 on compressor rotors for naval helicopters after ocean hovering. Thus, it is necessary to consider exposure to solid salt particles on titanium compressors and also on titanium airframes in transoceanic and naval service, which would of course include SST operation.

The possibility of hot salt corrosion of titanium was first encountered in laboratory tests from salt from fingerprints. It is now common to evaluate titanium alloys for susceptibility to hot salt stress corrosion by coating creep samples lightly with salt and checking residual ductility at room temperature after creep exposure.

The odd thing about hot salt stress corrosion is that it has never been encountered, or at least documented, in service. However, the conditions of stress, time, and exposure history rarely duplicate that encountered in laboratory tests. It is anticipated that the stress-temperature-time regime for supersonic transport airframes and engines may pose a threat for titanium service, and this is being carefully considered. A possible factor associated with the lack of correlation between service and laboratory experience is that cyclic temperature service is much less severe than steady loading for long times at temperature. Aircraft usage is of course cyclic with respect to airframe and engine temperatures.

Control

The conditions of operation of airframes and engines afford a most effective control against hot salt cracking. Hot salt cracking in the laboratory is rarely encountered below 500°F, which is above the anticipated maximum service temperature for supersonic transport operation at M 2.7. Also, it appears necessary to incur at least

0.1-1% plastic creep before hot salt stress cracking can be incurred. Thus, the creep limits for time, temperature, and stress are somewhat more severe than those for occurrence of hot salt stress corrosion. Lastly, under conditions of cyclic temperature loading with 1-2 hours at maximum temperature in laboratory creep tests, cracking has been inhibited under conditions that cause extensive cracking at the same time at maximum temperature when applied steadily.

All structural titanium alloys are susceptible to hot salt stress corrosion. However, several alloying elements appear to be detrimental, including high aluminum content. If the aluminum content is maintained at 6% or below, acceptable hot salt resistance is maintained. The transformed beta condition being more creep resistant would also be more resistant to hot salt stress corrosion, requiring a higher combination of time, stress, and temperature for initiation of salt cracking. There is also reason to believe that the finer grain sizes associated with transformed beta structures favor resistance.

Mechanism

The less severe effect of temperature cycling under stress appears to be associated with instability at low temperature of $TiCl_2$, one of the reaction products of hot salt stress corrosion. Thus, damage during the high temperature portion of a cyclic temperature is not cumulative. The reaction between NaCl and the TiO_2 scale on the alloy in the presence of oxygen appears to be involved in the hot salt stress corrosion reaction. Apparently, the salt becomes hydrolized and forms HCl, which penetrates the oxide film and reacts with the alloy forming hydrogen. The cracks formed at temperature reduce residual ductility, while the absorbed hydrogen further reduces ductility according to the usual hydrogen embrittlement mechanisms.

RED FUMING NITRIC ACID

Early in the evaluation of titanium as a corrosion-resistant material, it was found that titanium was particularly resistant to nitric acid. However, when the water content was reduced to the level found in red fuming nitric acid (RFNA), the titanium was susceptible to very rapid reaction, and has actually detonated in several cases.

Control

Titanium is not recommended as a container for red fuming nitric acid. A minimum content of about 2.5% H_2O is needed for immunity.

Mechanism

Apparently, the absence of H_2O renders the protective TiO_2 film on titanium incapable of being renewed once cracked after sufficient strain. This reaction is rather similar to the failure in N_2O_4, where under anhydrous conditions HNO_3 is formed.

HALOGENATED HYDROCARBONS

Trichloroethylene has been used as a degreasing agent for titanium sheet parts which were subsequently stress relieved. It was found that cracking occurred if the trichloroethylene was not completely removed, such as with titanium tanks containing trapped space. Other degreasing agents, such as Freon-TF, will cause stress corrosion cracking whereas Freon-MF will not.

Another instance of cracking by halogenated hydrocarbons occurred when a titanium alloy tank was pressure-tested at elevated temperatures using a high-boiling-point chlorinated hydrocarbon liquid (Aerochlor 50) as the pressurizing media. Also, cracking by temperature-indicating crayons made of halogenated hydrocarbons has been noted.

Control

In fabricating titanium parts, it has become standard practice to use "clean room" techniques, to avoid handling titanium parts by hand, and to remove oils or degreasants completely before subsequent heating. High temperature contact with halogenated hydrocarbons, which might dissociate and form HCl, must be avoided.

Mechanism

Apparently, dissociation of the halogenated hydrocarbons, forming HCl, which is very corrosive to titanium, occurs at elevated temperatures (about 600°F). Stress results in sufficient strain to break the oxide film, and expose the metal to HCL.

IGNITION AND BURNING

Titanium alloys have been considered for liquid oxygen (LOX) tankage, but have been found capable of being detonated when impacted in contact with LOX under pressure.

There have been about 30 titanium compressor in-flight fires in the UK, but very few in the US, although some "scorches" have occurred in test stand runs. Jet engine experts in the USA attribute the difference between UK and US experience to the British practices of allowing less tip clearance in the compressor and relying less on test stand work and more on flight test proving during development stages.

Control

Titanium is not satisfactory as a tankage for pressurized LOX. Studies on ignition proneness of various titanium alloys have not shown much difference. Compressor design to avoid Ti-Ti or Ti-steel rubs by permitting larger tip clearance or by spraying the titanium with a less-active material, like zirconia, will minimize this hazard.

LIQUID METALS

Titanium jet engine compressor parts have failed on test stands or in service as a result of reaction with molten metals or metal salts. One instance occurred from the inadvertent use of cadmium-plated steel bolts for a titanium alloy compressor in an experimental engine. After a hot spin test, it was found that cadmium penetrated the grain boundaries of the wheel, and failure occurred as a result of the reduction in the stress-bearing section of the wheel.

GE-Evandale encountered an instance of molten metal (or metal salt) failure as a result of the use of silver plating in the compressor slots to reduce fretting. In this case, it is believed that the silver formed AgCl by reaction with salt, and that molten AgCl caused the failure.

Control

Cadmium-plated bolts should not be used in contact with titanium exposed above the melting point of Cd (610°F).

In the case of silver used to guard against fretting, it has been possible to substitute organic-bonded dry lubricants, graphite grease, or molybdenum disulfide.

Mechanism

As is the case in many stress corrosion situations, when the stress is sufficient to crack the protective oxide surface film, there is danger from reaction of the environment with the fresh metal exposed at the base of the crack. If the environment is a molten metal or salt which reacts with the titanium, or reduces its interatomic bond strength, stress corrosion ensues. All molten metals do not cause stress corrosion of titanium, because they may not be sufficiently reactive. Thus, titanium resists molten alkali metals like sodium and potassium up to at least 1000°F.

SUMMARY AND RECOMMENDATION

As a result of recent aircraft and aerospace vehicle experience, titanium alloys have been found to be susceptible to many stress corrosion environments. In most cases the situation can be fixed by substitution of another titanium alloy, change of metallurgical condition, or addition of inhibitors to the environment.

It is recommended that before titanium is used in any new application it should be checked against susceptibility to stress corrosion in the particular environmental conditions involved in the proposed application.

INFLUENCE OF STRESS AND ENVIRONMENT ON THE STRESS-CORROSION CRACKING OF HIGH STRENGTH ALUMINUM ALLOYS

by

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INFLUENCE OF STRESS AND ENVIRONMENT ON THE STRESS-CORROSION CRACKING OF HIGH STRENGTH ALUMINUM ALLOYS

Robert H. Brown, Donald O. Sprowls and M. Byron Shumaker

1. CLASSIFICATION OF DELAYED CRACKING FAILURES

To define stress-corrosion cracking is a controversial task. The definition given in the Corrosion Handbook[1], "Cracking resulting from the combined effect of corrosion and stress", is a broad definition and actually covers a variety of chemical-mechanical failures. H.R. Copson[2] restricted the stress to "static tensile stress" and then classified stress-corrosion cracking failures into four types. However, the classification covers such a broad range of causes and mechanisms that stress-corrosion cracking has come to have different meanings to different people.

It is advantageous to have a classification of these varied failures to assist in cataloging service failures and in determining suitable preventative measures. This is a matter of great practical importance to materials engineers and designers because methods for avoiding each type of failure generally are different. A proposed classification based on the chemical-mechanical characteristics of the failures using generally accepted terms that have been used to describe the several types of failure is given in Table I.

Type 1 is stress-corrosion cracking and is defined as cracking that is initiated by directional chemical and/or electrochemical attack synergized by sustained tension stress at the surface. Type 2 is corrosion fatigue which is cracking that is initiated by cyclic stressing synergized by chemical and/or electrochemical attack. An important distinction between Types 1 and 2 in aluminum alloys is the mode of cracking: stress-corrosion cracks are characteristically intergranular, whereas corrosion fatigue cracks, even if they should initiate at a site of intergranular corrosion, are characteristically transgranular[3].

Type 3 is a mechanical type failure that sometimes is confused with true stress-corrosion cracking. This type of mechanical failure occurs only under severe corrosive conditions with the applied stress having no appreciable effect upon either the rate or the geometry of the corrosive attack. The distinction between these two types of failure is of considerable practical significance, however, because the methods of combating them are quite different. Whereas, the relatively simple procedure of applying a protective coating will suffice to prevent the corrosion that may cause a mechanical failure, the prevention of stress-corrosion cracking requires a much more sophisticated approach. Moreover, a corrosion induced mechanical failure may occur with alloys that are highly resistant to stress-corrosion cracking.

Types 4, 5, 6 and 7 will not be discussed since oxidation reduction of the parent metal is not directly involved in the cracking process. There may well be an electrochemical aspect associated with failures by hydrogen-embrittlement cracking.

2. FOUR AGENTS IN STRESS-CORROSION CRACKING

Aluminum alloys in tempers that are susceptible to stress-corrosion cracking are characterized by microstructures wherein there has been a localized decomposition of solid solution at the grain boundaries; in most instances, identifiable precipitation in the grain boundaries can be established. Stress-corrosion cracking occurs typically along the grain boundaries in contrast to the transgranular cracking generally associated with mechanical fractures resulting from fatigue, creep rupture, tensile overload, etc. Anistropy of the grain structure, influenced by the composition and by the conditions under which an aluminum alloy product is worked from the cast ingot, has a marked influence upon stress corrosion performance. Therefore, any alloying addition or metallurgical treatment that affects the precipitation of alloy constituents or the shape of the metal grains can markedly influence the resistance of an alloy to stress-corrosion cracking.

It cannot be overemphasized that the stress-corrosion cracking process requires not only a susceptible alloy but also a sustained tensile stress and exposure to particular and specific types of environment. Although an alloy of a given composition must have a specific microstructure in order that stress-corrosion cracking occur in a specific environment, an alloy of different composition but of similar microstructure will be resistant to stress-corrosion cracking. Moreover, the environment that causes stress-corrosion cracking of an alloy of a given composition and microstructure will not cause an alloy of different composition and same microstructure to stress corrosion crack. With such limitations the joint action of these agents not only increases the other factors' effectiveness in causing stress-corrosion cracking, but also, if one agent is absent, such as stress, stress-corrosion cracking does not occur.

The terms "susceptible alloy" or "susceptible microstructure" are meaningless to the intelligent use of high strength aluminum alloys. Figures 1 through 5 show electron micrographs from the work of Robinson, Lifka and Shumaker[4] illustrating the difficulty of trying to correlate structures with stress-corrosion resistance. The structures shown are for commercially fabricated forgings and one- and two-inch thick plate, and the stress-corrosion data are for short transverse tests. Structures of 7075 alloy in the W, T6 and T73 tempers are shown in Figures 1, 2 and 3, respectively. The structure representative of the maximum strength T6 temper differed from that of the W temper principally by the high density of zone formation in the T6. The T73 temper overaged in comparison with the T6 exhibited zones of larger size with greater interparticle spacing and some platelets of M′ or M precipitate. A few residual quenched in dislocations also are present. The T73 temper is virtually immune to stress-corrosion cracking in chloride environments, whereas parts in the T6 and W tempers are susceptible at low stresses when tested in the short transverse direction. Thus the type of structure shown in Figure 3 might be judged to be representative of a stress-corrosion resistant metallurgical condition. However, Figure 4 shows alloy 7079 in an experimental temper with a similar structure, yet it was quite susceptible to stress-corrosion cracking. Likewise, Figure 5 shows a similar structure in a Cu-free alloy, 7039, which is susceptible to stress-corrosion cracking at low stresses in the short transverse direction.

For obtaining the specific result of stress-corrosion cracking, four primary agents must be concurrently contributing:

(1) a specific environment

(2) a prolonged sustained tensile stress at exposed surfaces

(3) a specific alloy composition

(4) a unique microstructure derived by metallurgical processing.

One important factor of the performance of heat treated aluminum alloys with regard to stress-corrosion behavior is the direction of tensile stressing in relation to direction of working and grain flow. Figure 6 shows the three-dimensional microstructure of a heat treatable aluminum alloy plate[3]. The effects of directionality are most distinct and definite in heat treatable alloy plate. The direction parallel to the direction of rolling (working) has been termed longitudinal. The direction transverse (and perpendicular to the rolling direction) to the rolling and parallel to the rolling surface has been termed the long transverse. The direction normal to the rolled surface has been termed the short transverse.

The direction of working can be more complicated in other products. Although not typical, the macroetched transverse section of an 8in. x 8in. x 24in. hand forging of 7075-T6, shown in Figure 7, illustrates how complicated a structure can be produced. The variation in stress-corrosion resistance, as effected by long axis orientation of the test bar in relation to grain flow, is marked[3]. The high order of resistance with the long axis parallel to the grain flow, as determined by the 3½% NaCl alternate immersion test, is comparable to similarly oriented specimens taken from rolled plate. Further, the resistance in this direction relative to the grain flow is notably higher than in the normal direction.

Figure 8 demonstrates that if the high degree of resistance to stress-corrosion cracking in the longitudinal direction of a highly oriented grain flow is sacrificed in order to improve the resistance in the transverse directions by reducing the degree of longitudinal orientation, an improvement of the short transverse direction is small, whereas the threshold in the longitudinal direction is decreased markedly[4].

3. ENVIRONMENTAL FACTORS

There seems to be little controversy over the contention that stress-corrosion cracking is the result of the synergistic interaction of mechanical and oxidation-reduction phenomena. That is, in combination with the mechanical processes, there is a chemical reaction that occurs on a relatively small scale that results in the oxidation of the alloy along some highly localized path, and the reduction of an element or compound in the environment. Hence, in the absence of an oxidant, stress-corrosion cracking would not be anticipated; that is, if the environmental conditions were a vacuum or an inert gas, stress-corrosion should not occur.

Most normally encountered environments will contain water (H_2O) in the form of liquid or vapor; in this sense the environment will be aqueous. In the broadest sense it need not be, but could contain other reducible inorganic or organic compounds. However, since the nonaqueous condition would represent a condition not usually encountered, the discussion will be limited to aqueous containing environments. Figure 9 (Ref.5) can be used as an introduction to the effect of environment on stress-corrosion specimen life. Short transverse tensile bars 1/8in. in diameter from 3in. thick 7039-T651 plate were exposed, stressed to 75% of the short transverse yield strength. In chloride containing environments, stress-corrosion cracking occurred in less than a month. Although the specimens had a life of about 10 months in a mild laboratory atmosphere, the life was on the order of about 3 days in the New Kensington atmosphere which is normally a rather mild environment. The fact that these same type specimens stressed in a like manner have not cracked after having been stressed to 75% of the yield strength and immersed for six years in a highly refined mineral oil would seem to offer sufficient evidence that an oxidant is required to cause stress-corrosion cracking.

Shumaker has shown that C-ring specimens from 1½in. plate of the same alloy will stress-corrosion crack in distilled water when stressed to 35 ksi. Since tests made with C-ring specimens are considered somewhat less severe than when 1/8in. diameter bars are employed, it would appear this alloy is not highly stress-corrosion resistant.

Data such as shown in the previous figure have caused some to comment that since water alone can cause stress-corrosion cracking, water is a specific stress-corrosion cracking agent[4]. Such a generalization neglects the fact that many aluminum alloys do not stress-corrosion crack in distilled water. In addition some aluminum alloys will stress-corrosion crack in aqueous media, but a specific combination of additional ions or compounds must be present in an aqueous solution.

Figure 10 indicates the performance of five aluminum alloys stressed to 75% of their short transverse yield strengths and exposed to six aqueous solutions (pH 7). These data demonstrate that at least one out of three specimens of 7039-T63 failed in less than sixty days when exposed to one normal solution of NaCl, NaBr, NaI, NaF, $NaNO_3$ and $NaCrO_4$ (Ref.4). In contrast 7075-T651 failed only in Cl^- and Br^- solutions and the 2219-T37 failed only in the Cl^- solution. It is noteworthy that the 2219-T87 and the 7075-T73 did not stress-corrosion crack even after sixty days exposure in any solution. It is obvious from these data that the specific nature of the environment can vary with alloy composition and alloy temper. Further, an alloy such as 2219-T37 may be susceptible in chloride and water solution but not in nitrate and water solution.

This type of information becomes important in devising suitable stress-corrosion tests for evaluating the stress-corrosion susceptibility of alloy products under service conditions. In Figure 11 Shumaker[7] has shown that the relative susceptibility to stress-corrosion cracking of two tempers of 7039 in New Kensington atmosphere cannot be predicted by alternate immersion tests in NaCl solutions.

The stress-corrosion performance in an unusual environment emphasizes the significance of the specificity of the environment required in some instances to cause stress-corrosion cracking[8]. Figure 12 demonstrates a high resistance to stress-corrosion cracking of 7075-T73 in NaCl and the relatively low resistance of 2024-T3 in the same medium. However, there resulted a strikingly different behavior when the environment was inhibited red fuming nitric acid at 165°F. In this instance stress-corrosion cracking did not occur in the case of 2024-T3, whereas normally considered highly resistant 7075-T73 failed.

4. INFLUENCE OF STRESS (INTENSITY)

Most investigators agree that for stress-corrosion cracking to occur, tensile stresses at the surface of the alloy are required. Furthermore, most investigators agree that the tensile stresses must be sustained for relatively long periods of time. It is important to realize that these surface tensile stresses must be of such prolonged duration as to permit relatively slow chemical and/or electrochemical reactions to occur. Since in general terms the rate of growth of a purely mechanically produced fracture is much more rapid than rates of reaction between chemicals and alloys, it becomes almost obvious that the tensile stress must be of prolonged duration to permit the chemical and mechanical phenomena to produce their mutual interaction. The evidence requiring sustained tensile stresses at the surface was presented in two early papers[9,10].

For many years one of the most useful methods of describing the susceptibility (or the resistance) of an alloy and temper to stress-corrosion cracking has been in terms of a "threshold" stress. The "stress-corrosion threshold" usually is defined as the lowest value of sustained tension stress that will cause stress-corrosion cracking under given test conditions (the implication being that stresses below the threshold are safe stresses). The actual existence of a threshold stress for the stress-corrosion cracking of a highly susceptible material remains controversial because it is difficult to obtain experimental proof. However, it is unnecessary to resolve this point for the purpose of conducting stress-corrosion cracking tests to predict the serviceability of an alloy and to compare the resistance to stress-corrosion cracking of one alloy with another. Physical limitations of test equipment govern the accuracy with which a stress-corrosion cracking "threshold stress" can be determined; and the selection of test environment, type of test specimen, method of loading and duration of test determine the magnitude of the "threshold". Therefore, in dealing with threshold stresses, the specific test conditions must be associated with the threshold stress data.

The procedure for determining a threshold stress is illustrated in Figure 13 by the data obtained for the relatively new stress-corrosion resistant T76 (T76510, etc.) temper for 7075 and 7178 alloy products. Test specimens loaded to various stress levels are subjected to a suitable corrosion test, and a stress-corrosion cracking threshold stress, as indicated tentatively by the dashed line in Figure 13, is estimated. It has been established through many years of stress-corrosion testing 7075 and 7178 alloy products that the results of an 84-day alternate immersion test in 3.5% NaCl predict very reliably the performance to be expected after many years of exposure to a seacoast atmosphere[3].

In addition to the fact that sustained tensile stresses are required at the surface, the greater preponderance of data indicates that a certain tensile stress level must be exceeded if the stress-corrosion cracking of a given alloy composition with a given fabricating and thermal history is to stress-corrosion crack. Figure 14 compares the stress-corrosion cracking of 7075-T6510 (one of the older tempers for this alloy) with 7075-T76510 and 7178-T76510 (a recent temper development) extrusions stressed in the short transverse direction relative to grain flow and exposed to industrial and seacoast atmospheres. This information showed that in a large number of tests if a stress of about 7 ksi was exceeded in the short transverse direction, stress-corrosion cracking of 7075-T6 was likely to occur. However, a limited number of tests made on extrusions in a new temper of 7075 and a related alloy 7178 show that no stress-corrosion cracking occurred until a stress of 25 ksi was exceeded. It is also important to note that 7178-T76510 had a tensile strength almost the same as 7075-T651.

There is little data to support a generalized relationship between tensile strength and safe stress levels or threshold levels. Although Figure 15 does indicate that the stress-corrosion threshold in the longitudinal

direction at least approximates the tensile yield in this direction, certainly no relationship exists between these characteristics when data for the short transverse direction is examined.

Evaluation of the available data collected in Figure 16 comparing measurements of fracture toughness as determined by unit propagation energy or tear-strength/yield-strength ratio with stress-corrosion threshold level does not indicate a simple relationship between stress-corrosion cracking level and fracture toughness.

After a sharp notch resulting from a corrosion fissure or a minute stress-corrosion crack has initiated in a stressed specimen or component, there is developed at the tip of the fissure or crack a highly localized concentration of stress that is difficult to quantify with a high degree of accuracy. The magnitude of the concentrated stress can be large and becomes of more significance to the initiation and propagation of a stress-corrosion crack than the nominal stress. Thus, while the nominal stress may be related to the initiation of stress-corrosion and of stress-corrosion cracking, the concentrated stress at the tip of the incipient fissure or crack may relate more directly to the propagation of a crack. From the viewpoint of fracture mechanics, considerable significance is ascribed to the stress situation in the vicinity of a stress concentrator, and this stress situation generally is characterized by a "stress field intensity factor"[11,12]. This term, denoted K_I, is the quantity representing the combined effect of crack (fissure, or flaw) dimensions and nominal stress field which influences the behavior of the crack. If the initiation and propagation of a stress-corrosion crack is considered as a form of "subcritical flaw growth", then it is appropriate to inquire whether or not the mechanical aspects of stress-corrosion cracking are better characterized by the net section stress or the local stress at the tip of a fissure, as described by the stress intensity factor.

Because of the relatively long periods of time required (and often the uncertainty) to develop stress concentrators, many investigators have turned to the use of "precracked" specimens and the use of fracture mechanics analyses of the stress intensities[13,14]. The increase in fatigue crack growth rate because of exposure in a corrosive environment does not justify the use of precracked specimens analyzed by fracture mechanics techniques to predict stress-corrosion performance. Most studies in attempting to develop stress-corrosion evaluations of alloys incorporating stress intensity concepts have employed ferrous and titanium alloys.

There are several programs under way, especially at the Alcoa Research Laboratories, investigating the relation of threshold stress to stress intensity, as developed by modern fracture mechanics. Figure 17 shows graphically a theoretical relationship between stress intensity K_I versus time to fracture[15]. However, much experimental work must be done and several investigations are in progress to determine if alteration of K_I with time upon exposure in a given environment is the result of stress-corrosion cracking phenomena or other mechanisms inherent in the testing method.

The program at the Alcoa Research Laboratories evaluating the suitability of fracture mechanics tests to predict stress-corrosion behavior has just gotten under way. Precracked short transverse wedge-force loaded specimens of 7075-T651 and 2024-T351 plate were stressed by fixed deflection and were exposed to a number of environments. Figure 18 illustrates some of the difficulties that have been encountered[16]. The crack-length/time relationship shown graphically in Figure 18a was obtained on stressed specimens immersed in a solution 0.6 molar with NaCl and 0.03 molar with Na_2CrO_4. Figure 18b shows that, in the case of 7075-T651 calculation of K_I from the crack growth data was relatively consistent. However, the crack growth data for duplicate specimens of 2024-T351 was inconsistent. Calculation of stress intensity from the crack length indicated widely varying crack growth rates for the same stress intensity. Obviously, if these data are correct, some unrecognized phenomenon may be operating in addition to the modern fracture mechanics mechanism.

In most aluminum alloy products, the threshold stress is markedly influenced by the direction of stress relative to the grain structure. The degree of difference between the threshold stress in the three directions, that is, parallel to the longitudinal grain flow (longitudinal), transverse to the longitudinal grain flow (long transverse) and normal to the surface (short transverse), is to a great extent controlled by the morphology and shape of the grains which in turn is related to fabricating history and composition. Curiously enough, the ancillary elements often control the grain shape in heat treated wrought aluminum alloys. For example, in Figure 19 an aluminum-zinc-magnesium-copper alloy in three-inch thick plate has grains in the heat treated and artificially aged condition that are equiaxed[17]. The addition of 0.17% manganese does not appreciably change the grain shape. Figure 19 shows a relatively low resistance to stress-corrosion in all directions in the instance of both alloys in the T651 temper. However, the stress-corrosion performance in the longitudinal direction in the case of the T651 temper is materially improved by increasing the manganese to 0.53%. It will also be noted that increasing the manganese content caused the grains to be less equiaxed and more platelike. The addition of either 0.09% zirconium or 0.19% chromium caused even a more distinct platelike grain formation. Figure 19 shows that this change in grain shape resulted even in a further improvement in stress-corrosion performance of the T651 temper, not only in the longitudinal but also in the long transverse direction.

Figure 20 shows that the difference in stress-corrosion resistance between the three different directions in the five different alloys is materially reduced if the T73 type temper is used although the grain shape is the same as in the T651 temper[17]. It will be noted, however, that the improvement in stress corrosion resistance is accompanied by a substantial reduction in mechanical properties.

5. SERVICE PERFORMANCE

The sources of sustained tension stresses at the surface that cause stress-corrosion cracking in aerospace structures rarely result from operating loads that are anticipated in the design. The stresses anticipated by the designer are usually sufficiently low, and the loads resulting from operating conditions are usually of such short duration in aircraft that stress-corrosion cracking is not a result. Practically all difficulties encountered by stress-corrosion cracking result from stresses that are not anticipated or considered by the designer. Hence, the main problems created by stress-corrosion cracking result from tensile stresses that are introduced during fabrication of a part, during heat treatment of a part, or during the assembly of parts.

Forming, swaging, joggling and bending are fabrication operations that can leave residual tensile stresses on local areas of the alloy surface of sufficient intensity to permit stress-corrosion cracking to occur in a susceptible alloy if the service environment is of that particular specific composition as to permit stress-corrosion fissures to develop. An example of stress-corrosion cracking of a tube of 2024-T3 alloy that has been swaged at room temperature is shown in Figure 21. The problem could have been prevented by one of two ways. The swaged 2024-T3 tube could have been artificially aged to the T81 temper. Although the stresses would be slightly reduced by artificial aging, they would still remain at a substantial level. However, the T81 temper is resistant to stress-corrosion cracking in most natural or service environments. Another approach would have been to swage the 2024 tube in the O temper and heat treat after swaging to the T42 temper. Care must be exercised in the latter approach to be sure excessive grain growth does not occur on post-swaging heat treatment.

Residual stresses caused by quenching from the solution heat treating temperature can introduce sufficient tensile stresses in a part to result in stress-corrosion cracking. In many instances the portion of the part that is in tension is below the surface of the part during the quenching operation. However, machining operations can result in surfaces being exposed that have sufficiently high residual tension stresses to cause stress-corrosion cracking. This is especially true where the direction of the tension stress is normal to the direction of grain flow, for example, the parting plane region of a die forging. Figure 22 is a photograph of a machined forging of 7075-T6 alloy after exposure to the 3½% sodium chloride alternate immersion test. Note the stress-corrosion crack in the hollow machined boss. This crack is parallel to the metal flow lines of the parting plane and, hence is a short transverse failure. Since some reduction in mechanical properties was acceptable, the part was changed to 7075-T73. This alloy in this temper, having a high resistance to stress-corrosion cracking (that is, a high threshold stress level, even in the short transverse direction), did not suffer this type of failure, as illustrated in Figure 23. Other methods to control stress corrosion of such a part will be discussed later.

Assembly stresses, such as misalignments between holes of two parts joined by fasteners, interference fits caused by inadequate tolerance control, the use of tapered fasteners without sufficient dimensional control, etc., are examples of the manner in which residual tensile stresses can be introduced during shop assembly. Figure 24 shows a drawn tube of 7075-T6 alloy into the end of which solid aluminum alloy plugs were interference fitted. The forcing of the plugs created a circumferential sustained tensile stress in the wall of the 7075-T6 tube. Delayed cracking developed in the wall of the tube. A microscopic examination showed it to be a typical intergranular stress-corrosion crack. The use of an alloy having a higher threshold stress level in natural environments, such as 7075-T73 or 2024-T81, would have avoided the problem. In order to use 7075-T6 drawn tube, it was necessary to more carefully control the tolerances between the inside diameter of the tube and the outside of the plug in order that circumferential tensile stresses be held at a relatively low level.

6. MINIMIZING STRESS-CORROSION CRACKING DIFFICULTIES

Stress-corrosion cracking problems in aluminum alloys can be avoided by careful attention to the manufacturing methods employed, by the choice of alloys and tempers, and by the adherence to sound design and assembly practices that are based on an understanding of stress-corrosion facts and mechanisms. The reliability of this statement is proven by the widespread use of the high strength aluminum alloy 2024 in the T3 and T4 tempers. Although certain products of this alloy, for example thick plate in these tempers, are known to have an appreciable susceptibility to stress-corrosion cracking, they have been used for many years with only rare instances of stress-corrosion problems. Those relatively few stress-corrosion problems that have been encountered have resulted from a lack of appreciation of stress-corrosion knowledge and experience.

Table II, in outline form, lists the items that must be considered in stress-corrosion cracking and the steps that can be taken to avoid this type of failure.

As discussed previously, if the alloy and temper to be used are known to have an appreciable stress-corrosion cracking susceptibility, manufacturing methods, such as the swaging operation previously discussed, should be avoided. It is good practice to select stress relieved tempers wherever possible. Obviously the benefits of stress relieved tempers will be lost if procedures, such as tube sinking and power bending, are performed after stress relieving practices have been employed. Products such as extrusions, rolled rod, rolled bar, tube and plate are stretched (and hence straightened) in order to minimize residual stress gradients. Likewise, hand forgings and die forgings are cold pressed to minimize quenching stresses. Mild and less drastic quenches may be used under some circumstances to minimize residual quenching stresses, but such processes may introduce other problems and should not be used without very careful evaluation. Thermal stress relief treatments generally are not employed for heat treated aluminum alloys because they lower the

mechanical properties. Joggling, swaging or cold drawing will cause high residual tensile stresses. Such operations should be carried out prior to solution heat treatment. (The latter relaxes the residual stresses.) If because of certain metallurgical problems this approach is not feasible and the material must be processed in the solution heat treated condition, the product should be artificially aged to a more stress-corrosion resistant temper, such as T81 or T73. Misfits, as when fastener holes in two mating parts do not match, or interference fits created by bushing components, can be a source of high sustained surface tensile stresses. If service conditions anticipate the need for a particular set of mechanical properties, and these can only be attained with an alloy of low stress-corrosion resistance, better fitting up must be achieved (for instance, by reaming the fastener holes) or better control of tolerances in the interference fits.

Machining of the external surface of parts that were not stress relieved will cause readjustment of the stress distribution, thereby risking the exposure of surfaces that are stressed in tension. Surface rolling or shot peening to introduce compressive stresses in the surface layers has been found to be beneficial. Such prestressing methods are usually employed when design or service conditions will not permit the choice of a more stress-corrosion resistant alloy and temper. However, in many instances, reconsideration of the design to use more stress-corrosion resistant alloys will result in a more satisfactory product with little or no weight penalty.

The use of such stress-corrosion resistant alloys is usually dictated if sustained tension stresses normal to the direction of grain flow cannot be avoided. For example, wrought alloys such as 2014-T6, 2024-T4, 7075-T6 and 7079-T6 have a high resistance to stress corrosion cracking in the longitudinal direction but a relatively low resistance if the stresses are applied in the short transverse direction. If short transverse stresses at the surface are to be encountered, the alloys and tempers such as 7075-T76, 7075-T73, 7178-T76, 2219-T81, 2219-T87, 2024-T72 and 2024-T81 should be given serious consideration by the designer. The stressing in the short transverse direction can often be avoided by the selection of a more suitable product, for example, by using die forgings. Stressing in the short transverse direction can often be avoided, whereas machining the part from a hand forging or extrusion may require stressing normal to the grain flow direction.

The depth to which compressive stresses can be introduced by surface rolling or peening is limited. Therefore, mild general surface corrosion may eliminate the compressive layer and its benefits are lost. Therefore, organic coatings, in addition to delaying stress-corrosion cracking, can be beneficial in preventing general surface corrosion that could result in the exposure of metal stressed in tension. The beneficial effects of painting and peening are shown in Figure 25 (Ref. 18).

An efficient method of preventing stress-corrosion cracking is the use of alclad products. Since alclad products consist of a high strength aluminum alloy core metallurgically bonded to a surface layer of another aluminum alloy that will electrochemically (cathodic) protect the core, it is obvious the surface layer should not be removed by machining. Most widely used alclad products are in the form of sheet and plate with cladding on one or both sides[19]. Other products of irregular contour, such as forgings, can have surface layers of a more anodic potential than that of the alloy applied by hot dipping, by electroplating, and by metallizing processes. Coatings, such as those applied by electroplating and metallizing, have proven to be the most beneficial and practical.

Obviously, an alloy of such composition that it can be processed to obtain a high threshold stress with high mechanical properties represents a superior solution. Figure 26 shows a brief look to the future[20]. The T76 temper of alloys 7075 and 7178 was developed to improve the threshold level over that of T6 temper and with mechanical properties superior to those of 7075-T73. Today plate thicknesses up to 1in. and extrusions with thickness portions not more than 1in. are available in this temper. It is believed that a stress-corrosion threshold stress for natural environments of not less than 25 ksi in the short transverse direction can be safely anticipated with a high resistance to exfoliation. Figure 26 compares the stress corrosion performance in an industrial atmosphere for 12 months.

Several experimental alloys of the 7XXX type are showing considerable promise of improved stress-corrosion performance. The results of 12 months exposure in an industrial atmosphere compares in Figure 26 specimens from one of these experimental alloys in the form of 3in. × 12in. × 9in. hand forging with those from a duplicate forging in 7175 alloy. Based on 12 months exposure, their stress-corrosion cracking performance is superior to that of 7175 and with mechanical properties at least equal to those of 7075-T6. However, longer exposure with additional lots of material is required to substantiate present indications.

7. SUMMARY

1. Stress-corrosion cracking is defined in this paper as cracking that is initiated by directional chemical and/or electrochemical attack synergized by sustained tension stress at the surface.

2. Stress-corrosion cracking results when the following four agents are present simultaneously:

   (a) an enduring tensile stress at exposed surfaces
   (b) a specific composition
   (c) a unique microstructure, and
   (d) a corrosive environment of a specific nature.

3. Sustained tensile stresses at the exposed surfaces are required for stress-corrosion cracking to occur. The relationship of a critical stress level to stress intensity, as developed by modern fracture mechanics techniques, is under current investigation as part of an evaluation of tests to determine the suitability of fracture mechanics techniques for predicting stress-corrosion cracking performance.

4. Service failures of aluminum alloys by stress-corrosion cracking have been very infrequent. These are generally caused by sustained surface tensile stresses introduced by certain fabrication or assembly procedures. Also, relieval of surface stresses by peening or rolling may be effective. If it is not possible to avoid these stresses through control of fabrication and design practices, a change of alloy or temper is recommended.

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TABLE I

A Classification of Delayed Cracking Failures

Cause of cracking	*A* *Stress*	*B* *Oxidation-reduction*	*Interaction A and B*	*Type of failure*
Stress Corrosion	Sustained Tension	Yes	Synergistic	Chemical Mechanical
Corrosion Fatigue	Intermittent	Yes	Synergistic	Chemical Mechanical
Area Reduction by Corrosion	Sustained Tension	Yes	Additive	Mechanical
Hydrogen Embrittlement	Sustained Tension	No	None	Mechanical
Fatigue	Intermittent	No	None	Mechanical
Creep Cracking	Sustained Tension	No	None	Mechanical
Liquid Metal Penetration	Sustained Tension	No	None	Mechanical

TABLE II

Minimizing of Stress-Corrosion Cracking

I	Minimize the Magnitude of Residual Tension Stresses
	(a) Manufacturing Methods
	(b) Quenching
	(c) Design - Interference Fits
	(d) Assembly - Misfits
II	Stress Relief
	(a) Thermal Treatments
	(b) Introduce Surface Compressive Stress - Peening
III	Selection of Resistant Material
	(a) Composition
	(b) Product
	(c) Metallurgical Condition
IV	Organic Coatings
V	Cathodic Protection
	(a) Clad Alloys
	(b) Coatings - Hot Dip, Electroplate, Metallized

Fig. 1 Electron transmission micrograph of thin foil prepared from rolled plate of 7075-W heat treated in the laboratory and quenched in cold water. Dislocations are pinned to some particles of chromium-rich constituent (C) with no visible evidence of zones in the matrix or of M′ phase precipitate in the grain boundaries. Short transverse specimens were susceptible to stress-corrosion cracking at 25% Y.S.

Fig. 2 Electron transmission micrograph of thin foil prepared from rolled plate of 7075-T6 aged from the 7075-W plate shown in Figure 1. Similar to that of the W temper except that minute zones are visible in the grains (black specks Z) and numerous particles (P) are present in the grain boundaries. Dislocations are pinned to some particles of chromium-rich constituent (C) and to some particles of grain boundary precipitate (P). Short transverse specimens were susceptible to stress-corrosion cracking at 25% Y.S.

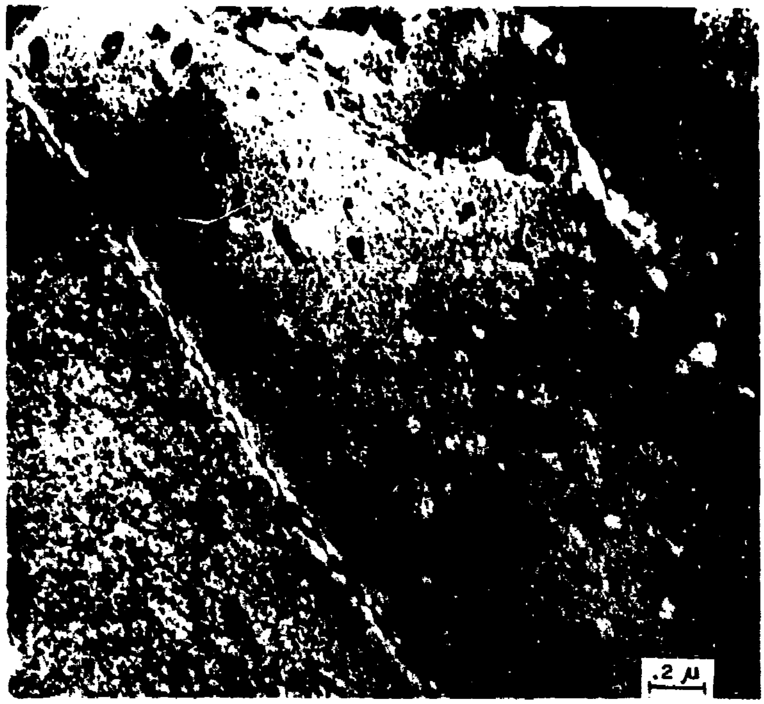


Fig.3 Electron transmission micrograph of thin foil of 7075-T73 aged from the 7075-W plate (Fig. 1). When compared with 7075-T6 (Fig. 2), showed the zones to be larger and the spacing between them to be greater. Tiny platelets of M′ precipitate (M′) also are visible along with residual quenching dislocations (D). An increase in the density of the grain boundary precipitate is evident with a very narrow region devoid of zones adjacent to the boundary precipitate. Short transverse specimens were resistant to stress-corrosion cracking at 75% Y.S.

Fig 4 Electron transmission micrograph of thin foil prepared from a forging of 7079 alloy in an experimental temper. While the microstructure is similar to that of 7075-T73 (Fig.3), short transverse specimens were susceptible to stress-corrosion cracking at 50% Y.S.

Fig. 5 Electron transmission micrograph of thin foil prepared from a rolled plate of 7039 alloy in a T6 type temper. The microstructure is not distinguishable from that of 7075-T73. Nonetheless the short transverse specimens were susceptible to stress-corrosion cracking at 25% Y.S.

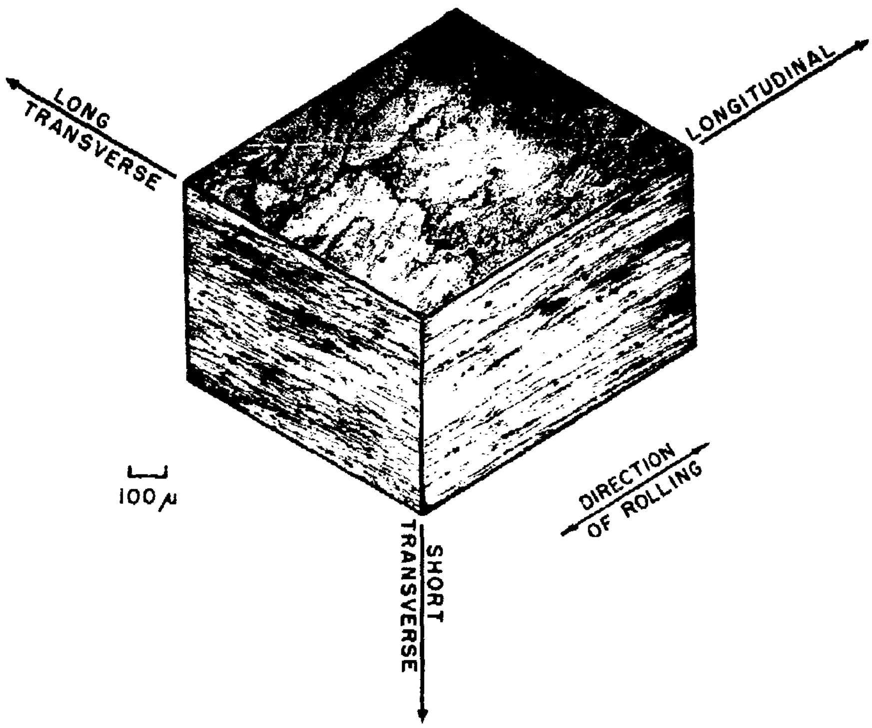


Fig Composite micrograph illustrating the grain structure of 1½in. thick plate of 7075-T6 alloy. The differences in grain shape of plate in three dimensions is more definitive than in most other wrought products. The relatively long, wide and thin unrecrystallized fragmented grains are typical of the grain structure of thick plate of other alloys also.
(Etch: Keller's; magnified 100X)

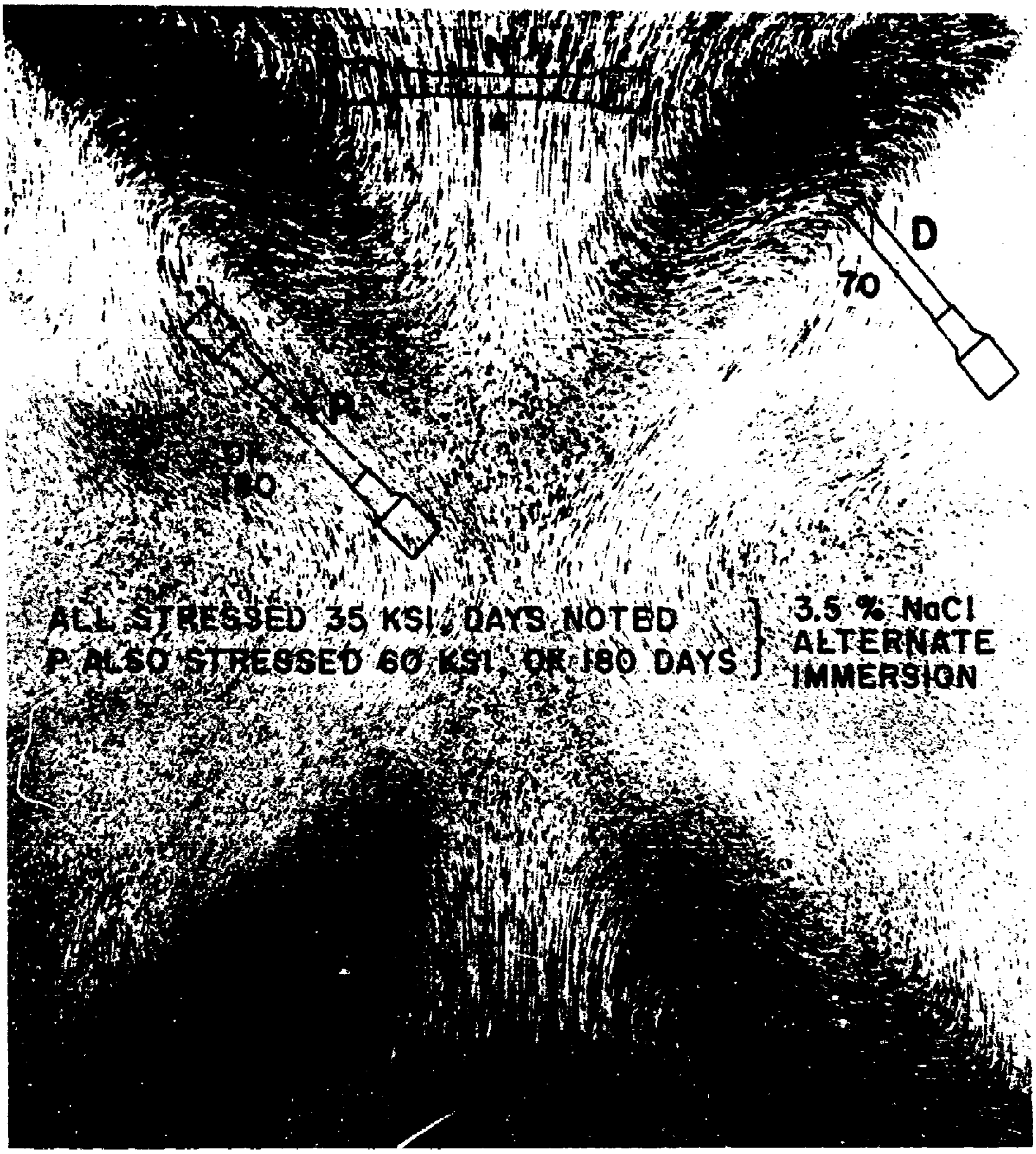


Fig. 7 Superimposed on a photograph (1X) of a macroetched transverse section of a special 8in x 8in. x 24in. hand forging of 7075-T6, fabricated so as to produce a complex grain flow, are the outlines of three stress-corrosion specimens along with their days to failure in the 3.5% NaCl alternate immersion. These specimens showed widely different stress-corrosion resistance, as would be expected in view of their orientation to the grain structure. Noteworthy is the high order of resistance of specimen (P).

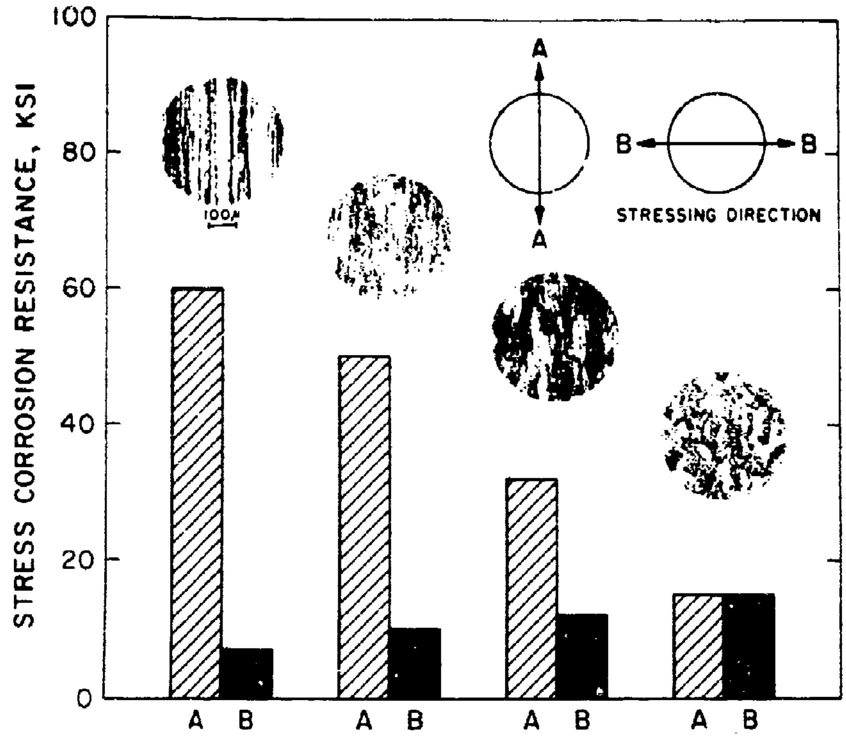


Fig. 8 Stress-corrosion cracking threshold stress of 7075-T6 extrusions of varying degree of grain geometry (degree of grain orientation) was determined in two directions, that is, parallel to, and perpendicular to the principal metal flow. It is evident that as the degree of directionality decreases, the increase in stress-corrosion resistance in the perpendicular direction (short transverse or transverse) is much less than the decrease in resistance in the parallel direction (longitudinal).

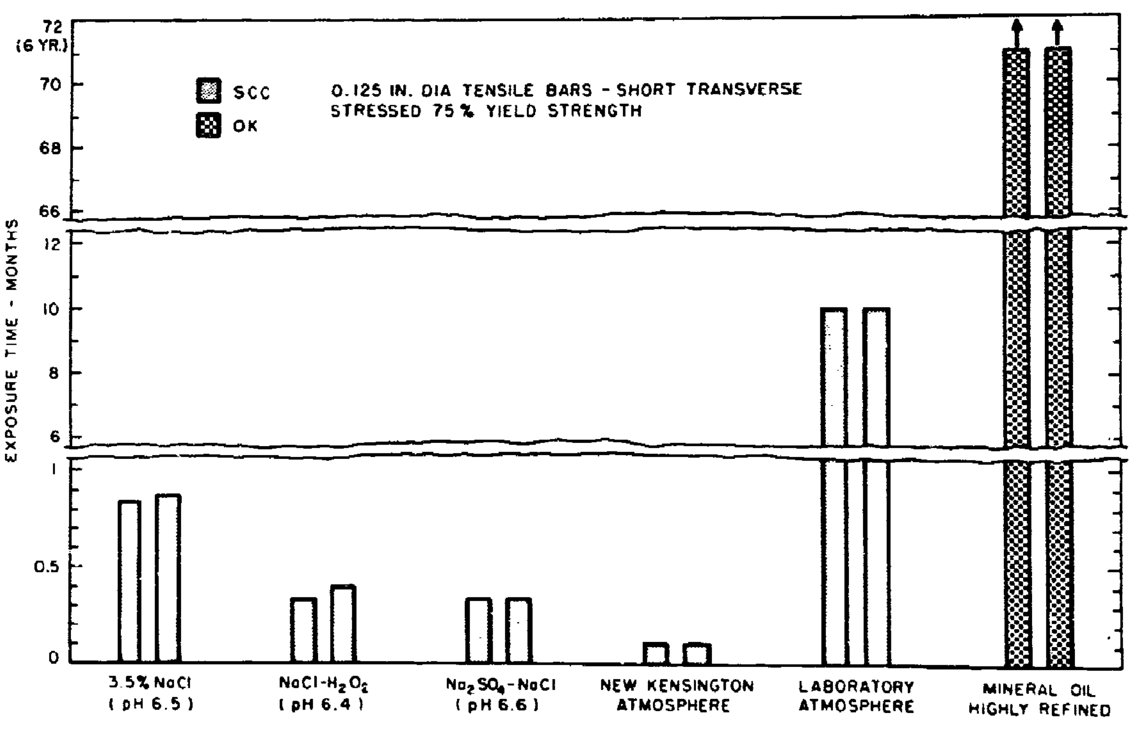


Fig. 9 The effectiveness of several environments in causing stress-corrosion cracking of short transverse 1/8in. diameter tensile specimens from 3in. thick 7039-T651 alloy plate. Bars indicate individual test specimens

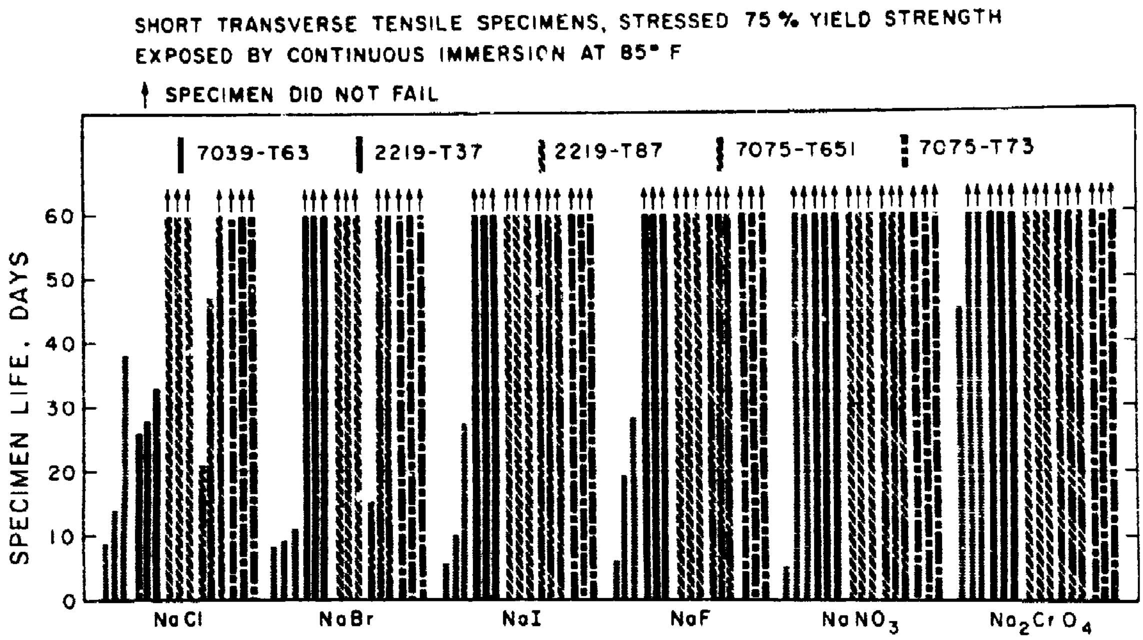


Fig. 10 The efficacy of several one normal aqueous solutions of sodium salts in causing the stress-corrosion cracking of short transverse 1/8in. diameter tensile specimens from 2in. thick plates of five heat treatable aluminum alloys. Bars indicate individual test specimens

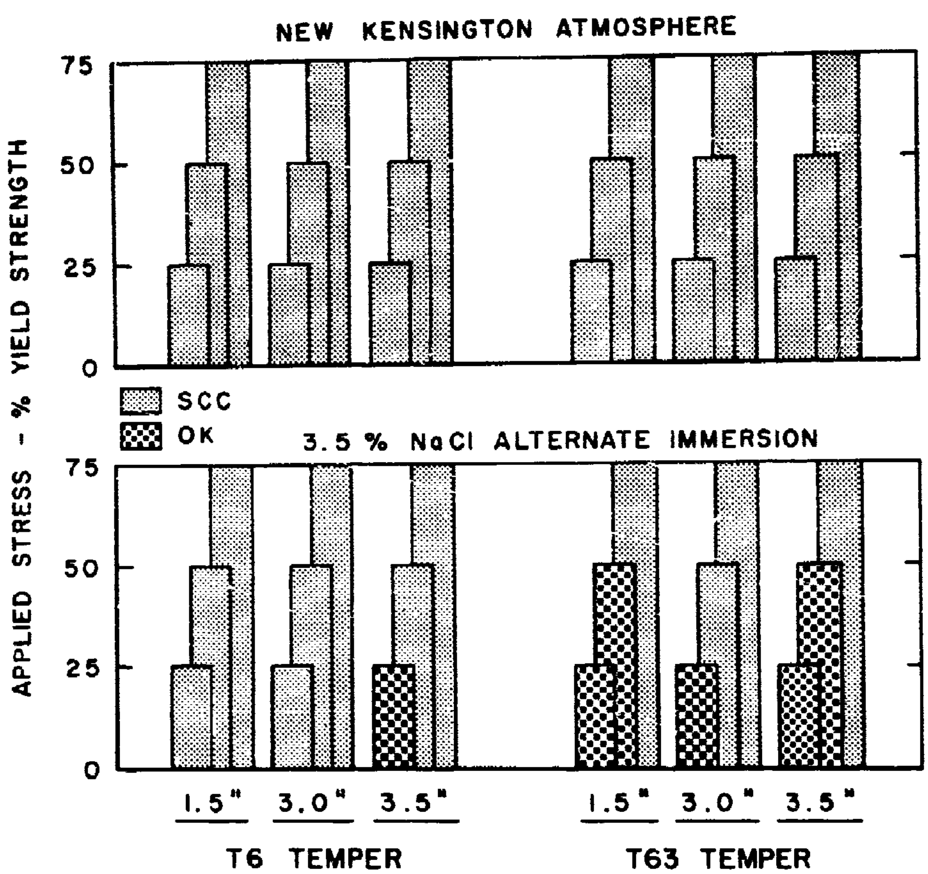


Fig. 11 Influence of corrosive environment on the stress-corrosion performance of short transverse specimens removed from the same lot of alloy plate 7039 in two tempers, namely T6 and T63

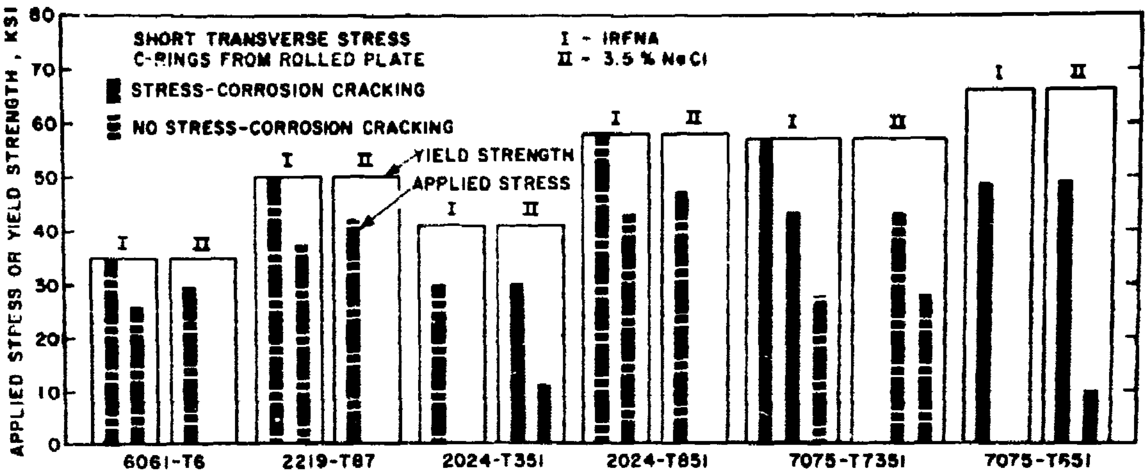


Fig. 12 Comparison of resistance to stress-corrosion cracking of various alloys in inhibited red fuming nitric acid at 165°F and in 3.5% NaCl alternate immersion. Bars indicate individual test specimens

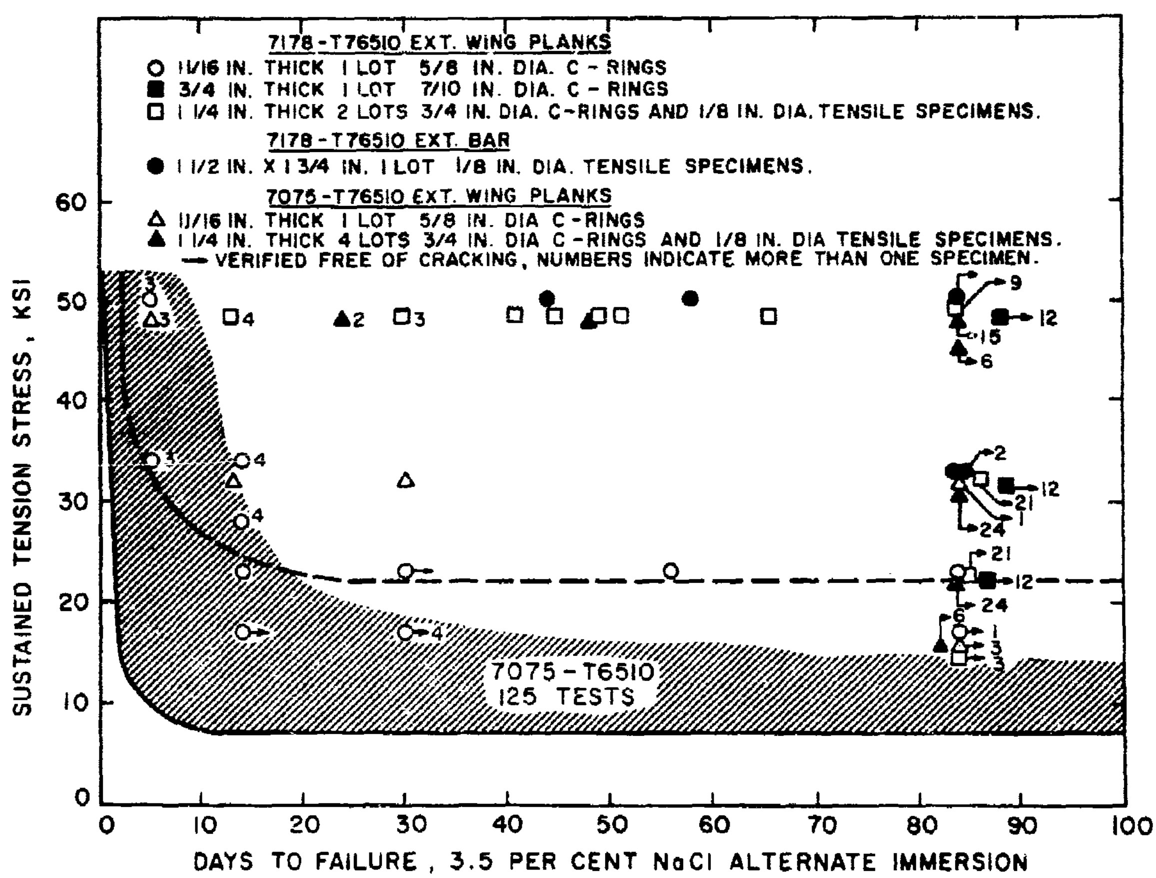


Fig. 13 Relative resistance to stress-corrosion cracking of high strength aluminum alloy extrusions tested in the short transverse direction in the 3½% NaCl alternate immersion. Specific test data for the relatively new stress-corrosion resistant T76 (T76510, etc.) temper for 7075 and 7178 alloy extrusions have been plotted on the background of test data for established materials of 7075-T610. The threshold stress is indicated tentatively by the dashed line

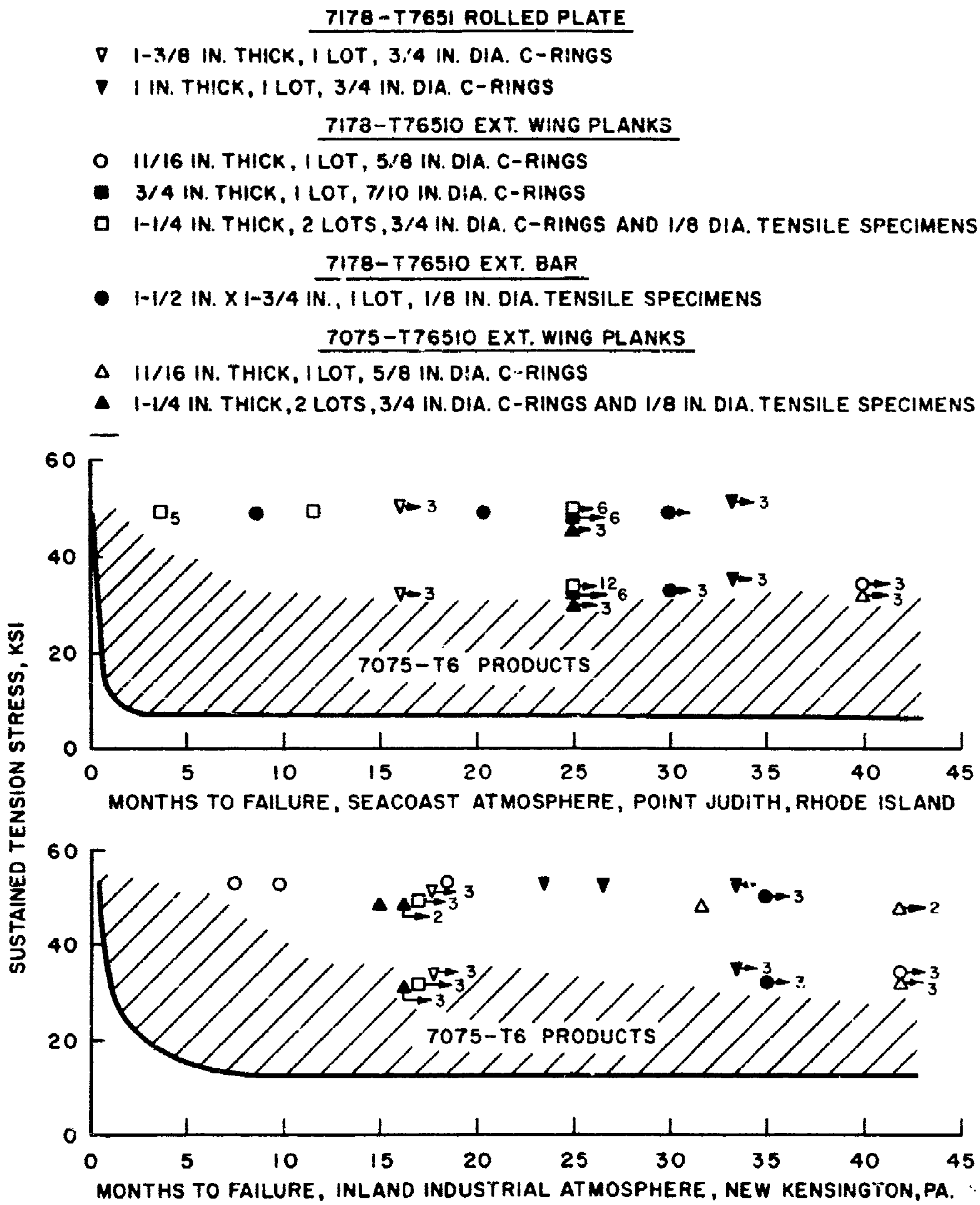


Fig. 14 Relative resistance to stress corrosion cracking of high strength aluminum alloy rolled plate and extruded products tested in the short transverse direction in an industrial and seacoast atmosphere. Specific test data for the relatively new stress-corrosion resistant T76 (T76510, etc.) temper for 7075 and 7178 rolled plate and extruded products have been plotted on the background of test data for established 7075-T6 products

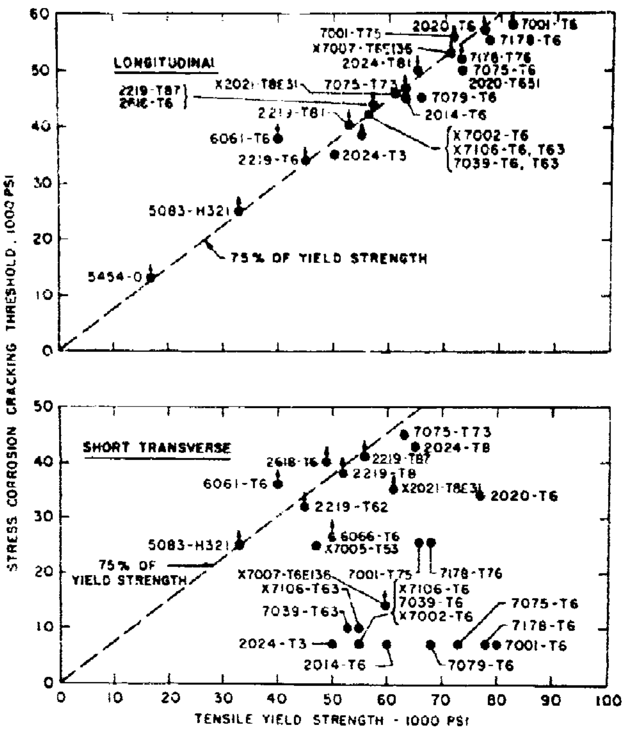


Fig. 15 Relationship of resistance to stress corrosion cracking with yield strength of aluminum alloy plate. It would appear that the stress corrosion cracking threshold stress level for longitudinal test specimens of established alloys and tempers increases with increasing tensile yield strength. However, the correlation in the short transverse direction is extremely poor

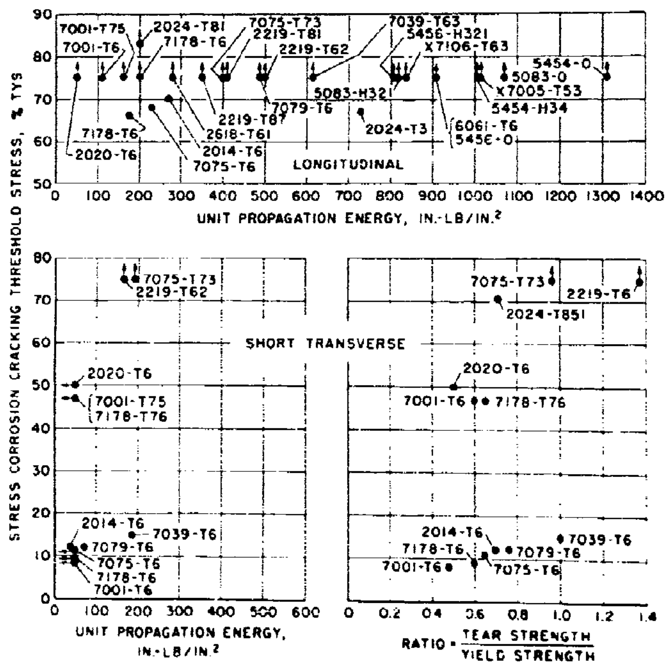


Fig. 16 These charts demonstrate the lack of correlation between the stress-corrosion threshold level in % of tensile yield strength obtained on smooth 1 8in. diameter tensile specimens and fracture toughness in terms of unit propagation energy (inch pounds per square inch) The correlation between stress-corrosion resistance (threshold in % of T Y S) and a measure of ductility, such as ratio of tear strength to yield strength, does not appear to exist

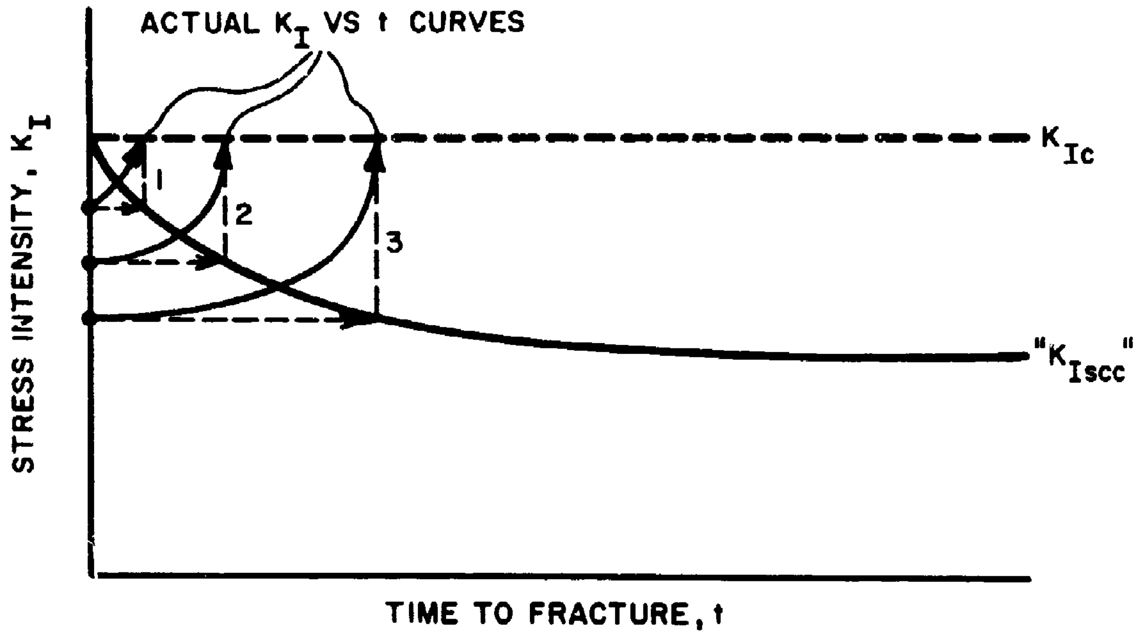


Fig. 17 Schematic illustration of probable stress intensity factor K_I versus time for an alloy susceptible to stress-corrosion cracking

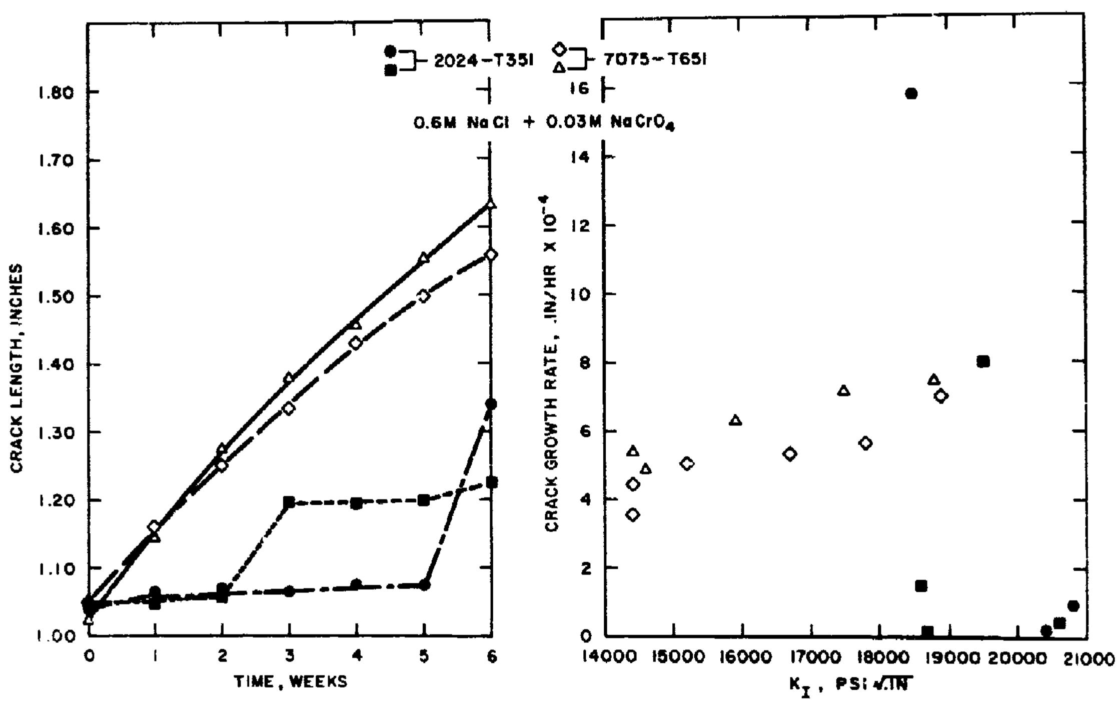


Fig. 18 Precracked short transverse specimens of 7075-T651 and 2024-T351 plate stressed by constant deflection and immersed in a solution of 0.6 Molar NaCl and 0.03 Molar $NaCrO_4$. Each kind of point indicates individual test specimens

(a) Crack length versus time

(b) Crack growth rate versus stress intensity factor K_I (pounds per square inch per square root of inches)

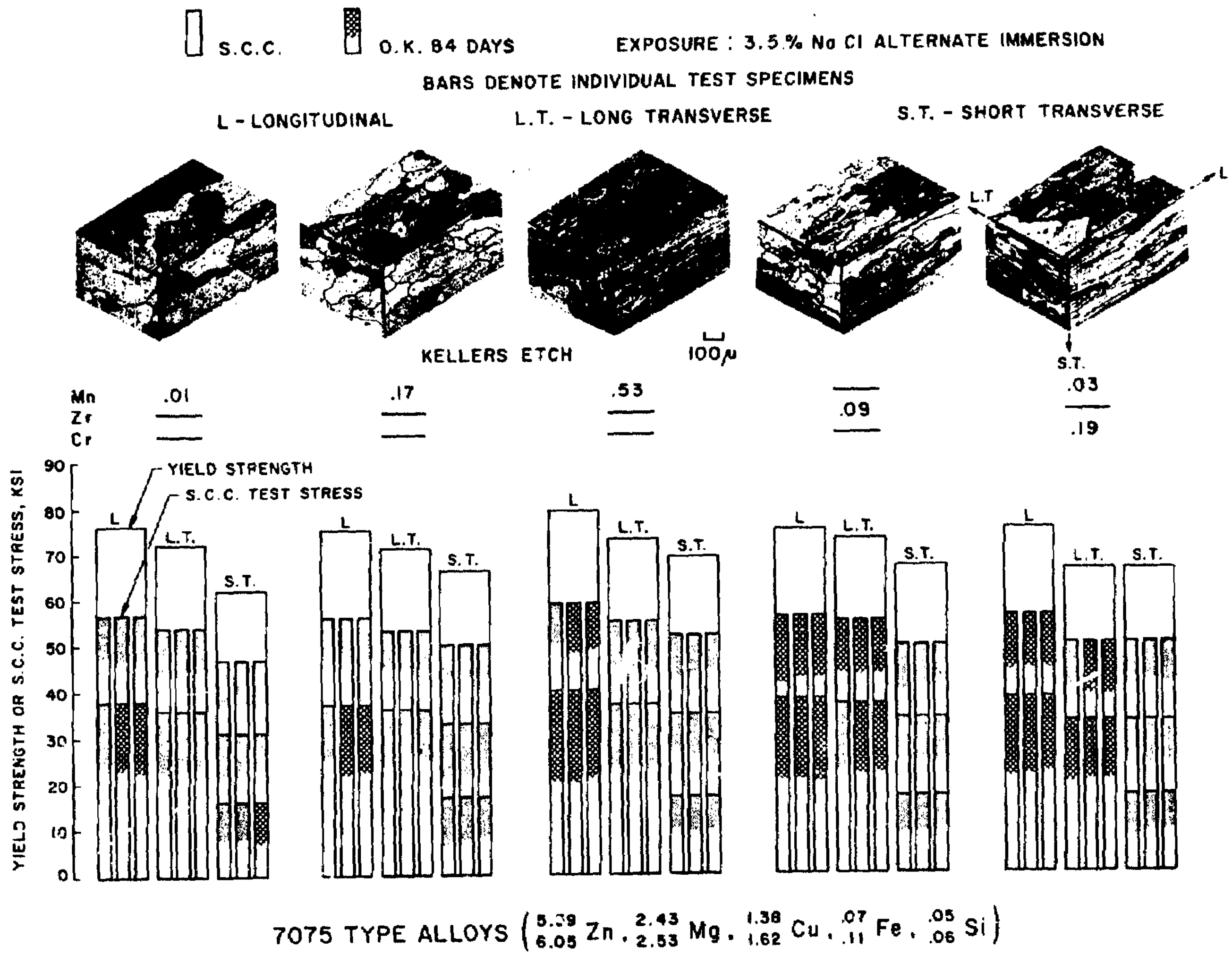


Fig. 19 Effect of grain shape of 3in thick plate of five 7XXX alloys (Zn 5.6 to 6.0% Zn, Mg 2.4 to 2.6% Mg and 1.4 to 1.6% Cu) with varying ancillary elements on stress corrosion performance of the T651 temper in 3.5% NaCl alternate immersion. Note that elongated plate-like grains improve stress corrosion performance in longitudinal and long transverse directions

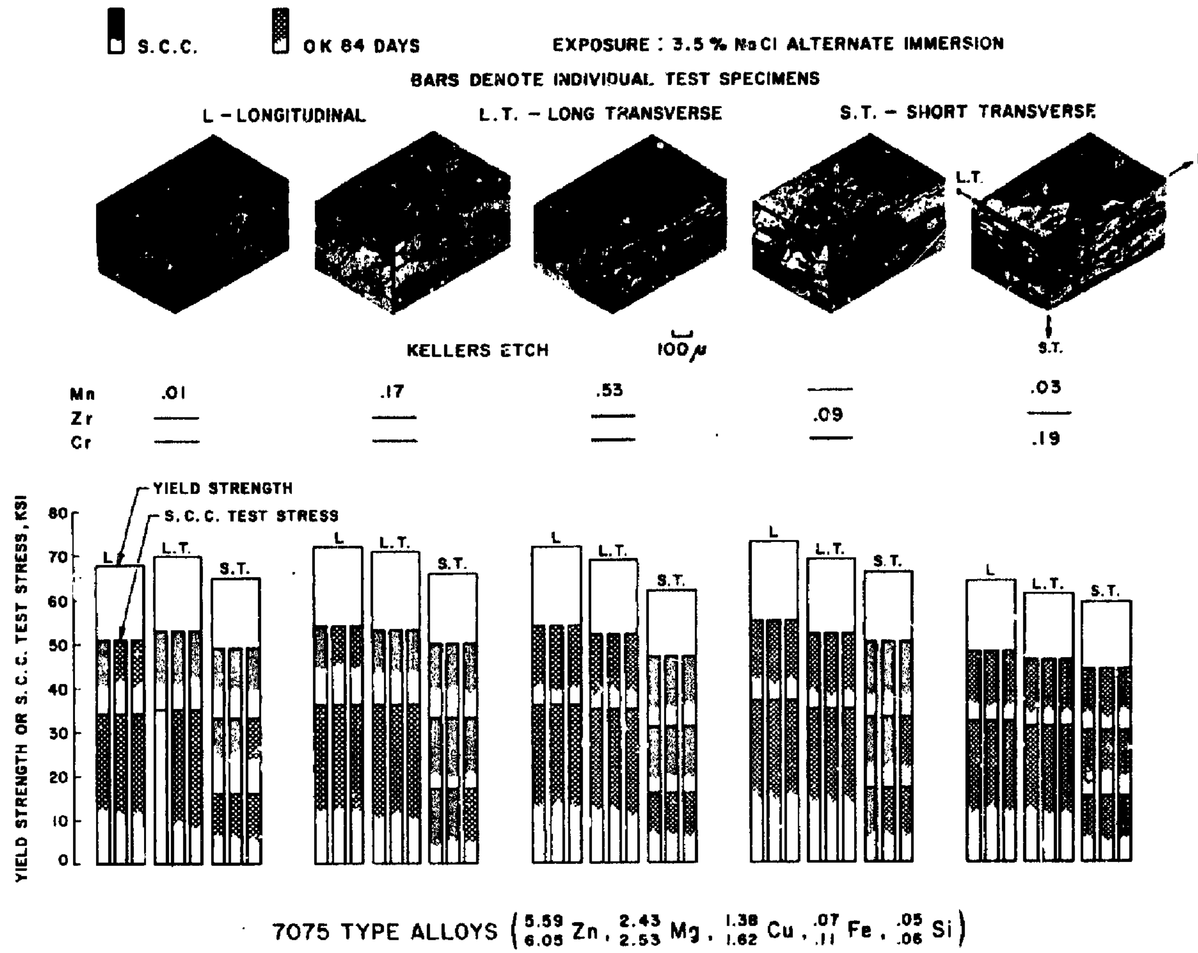
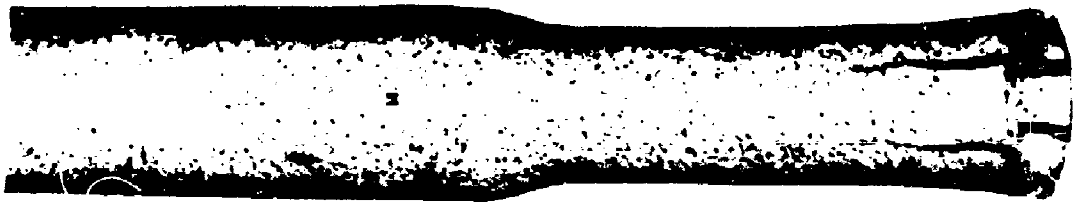
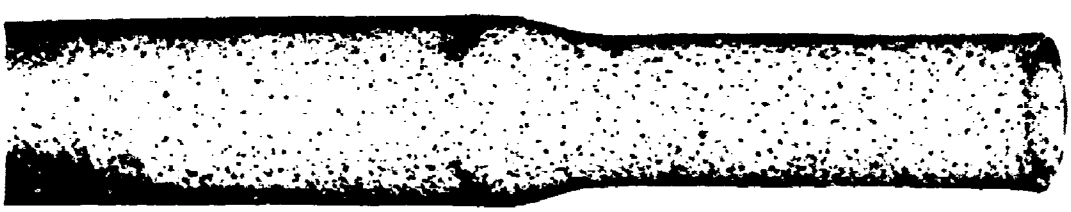


Fig. 20 Effect of grain shape of 3in. thick plate of five 7XXX alloy of same composition as listed for Figure 19 on the stress corrosion performance of the T7351 temper in 3.5% alternate immersion. Although there is an improvement of the T7351 over the T651 temper in stress corrosion resistance of all give alloys and the difference in three directions materially reduced, the changes achieved by T7351 in stress-corrosion cracking resistance are accompanied by an appreciable decrease in mechanical properties from those of T651 temper



SWAGED 2024-T3

SWAGED 2024-T3, AGED TO T81



SWAGED 2024-O, H.T. TO T42 1X

Fig. 21 Tubes with swaged ends of various tempers of 2024 alloy exposed to 3½% NaCl solution by alternate immersion for 84 days. (Corrosion products chemically removed after exposure.)

Fig. 22 Machined forging of 7075-T6 alloy showing extent of cracking into forward boss after an exposure of 18 days in the 3½% NaCl solution by alternate immersion. This crack is parallel to the metal flow lines of the parting plane and hence is a short transverse failure (Magnified approximately 1X)

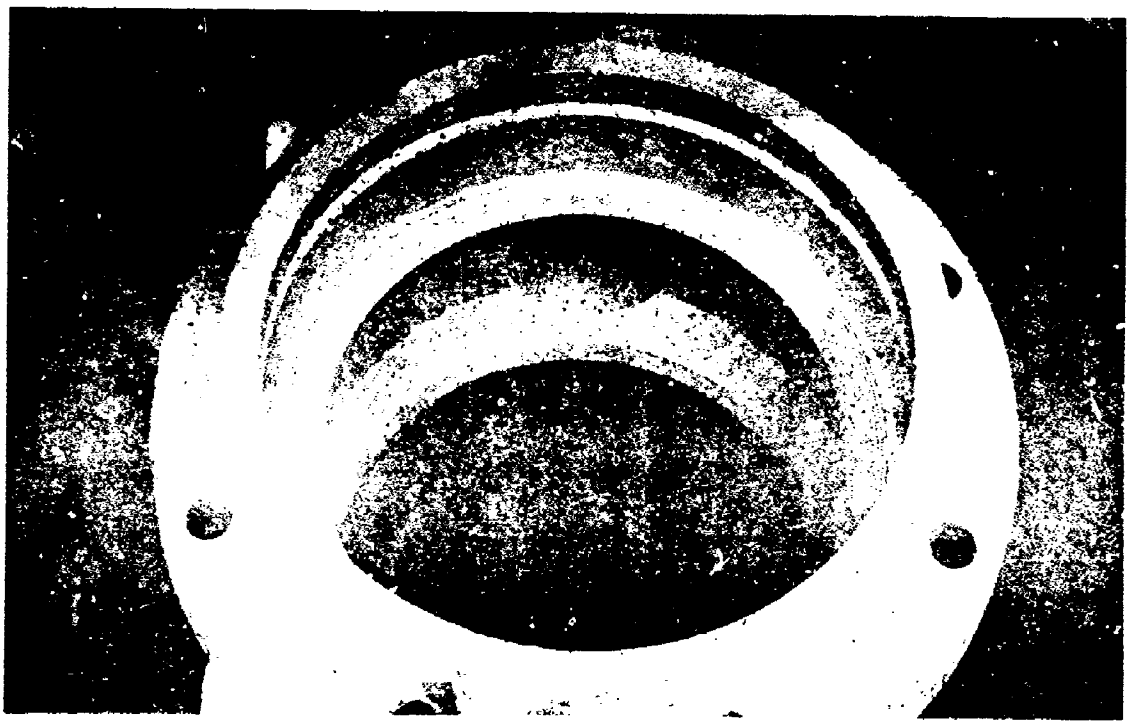


Fig. 23 Machined forging of 7075-T73 alloy identical to the 7075-T6 in Figure 22 showed no evidence of cracking after an exposure of 30 days in the 3½% NaCl solution by alternate immersion (Magnified approximately 1X)

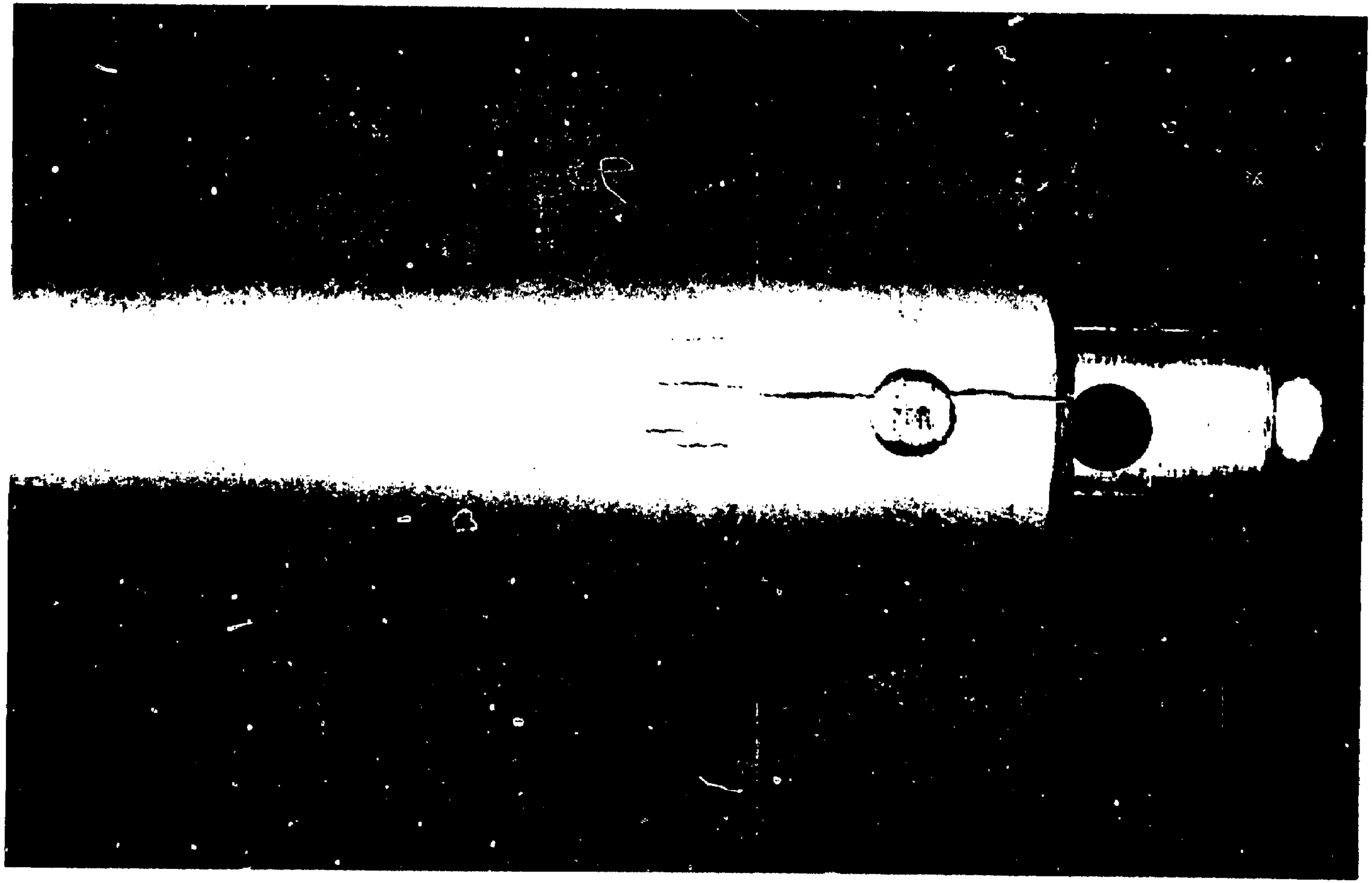


Fig. 24 Anodized ¾in. O.D. x 0.058in. wall 7075-T6 drawn tube containing pressed-in plugs failed in service. Cracked 7075-T6 tube end showing partially withdrawn plug

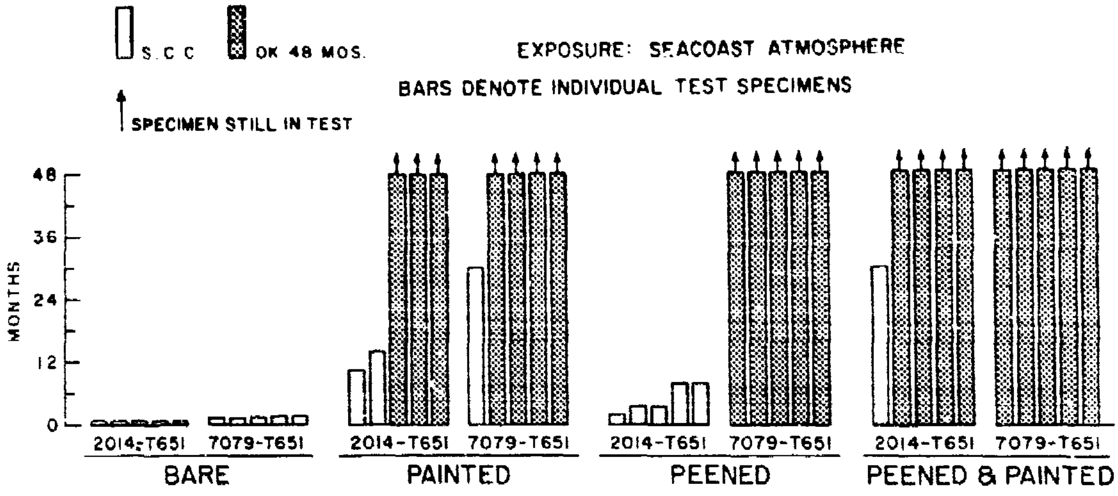


Fig. 25 Interference-fit rings (2¾in. diameter) from 2014-T6 and 7075-T6 alloy rolled rod peened prior to stressing to 75% of their transverse tensile yield strength. Painted specimens were coated with epoxy paint after stressing. Quintuplicate rings were exposed to an industrial atmosphere and seacoast atmosphere

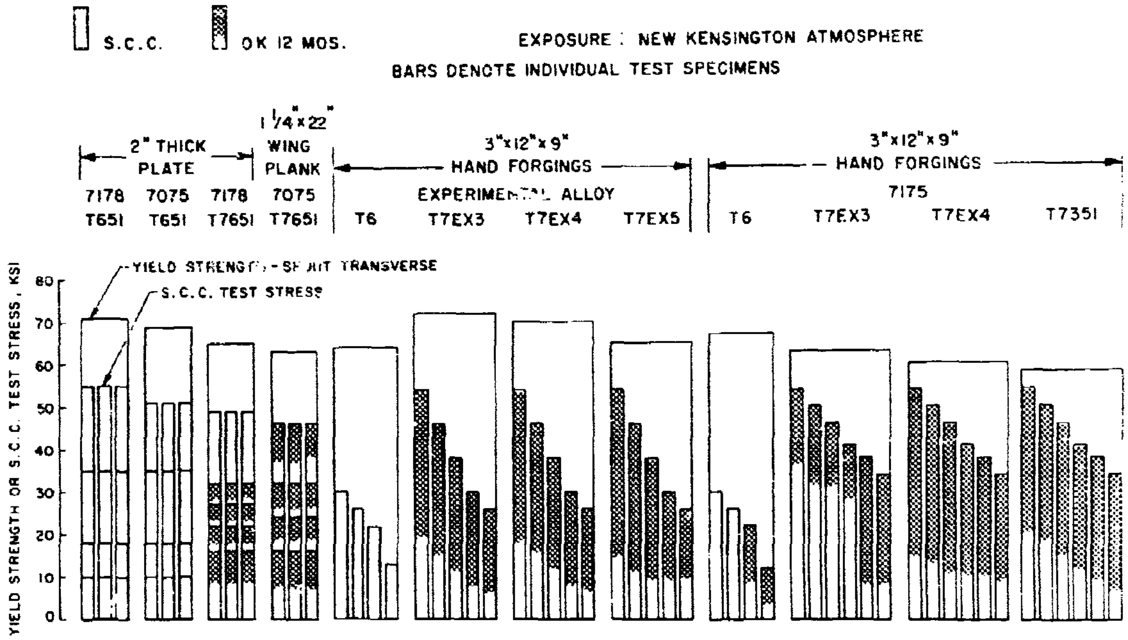


Fig. 26 This bar chart illustrates preliminary stress-corrosion cracking results (one year exposure in an industrial atmosphere) of recent alloy and temper developments in the 7XXX alloys

CHARACTERISTICS OF ENVIRONMENTALLY-INDUCED DELAYED FAILURE IN HIGH STRENGTH STEELS

by

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ABSTRACT

The delayed failure characteristics of high strength steels exposed to aqueous environments were reviewed with respect to the influence of both test conditions and material variables. Particular emphasis was placed on the results obtained from precracked specimens using linear elastic fracture mechanics to define crack growth rates and the stress intensity parameter below which cracking was not observed (K_{Iscc}). The addition of chlorides to water had only a relatively slight effect on delayed failure. The solution pH over the range of 3 to 10 also did not alter the environmental cracking characteristics. Dry hydrogen at one atmosphere was a particularly detrimental environment and produced crack growth rates that were greater than those obtained in moist air. In liquid environments increasing temperature produced an increase in crack growth rate with an activation energy equal to that observed for hydrogen embrittlement.

For a given steel type the K_{Iscc} value increased with decreasing tensile strength level. At a given strength, however, the K_{Iscc} was relatively independent of composition for the martensitic type, high strength steels hardened by carbon. The kinetics of the delayed failure process were significantly altered (approximately three orders of magnitude) by variation in the steel composition. The high nickel maraging steels had the greatest resistance to environmental cracking. Metallurgical structure also exerted a significant effect on delayed failure characteristics. Low carbon martensite or bainite in which microtwinning was absent had greater resistance to environmentally-induced delayed failure. The available data were consistent with a mechanism which involved embrittlement of the steel by hydrogen produced by a metal-environment reaction.

CHARACTERISTICS OF ENVIRONMENTALLY-INDUCED DELAYED FAILURE IN HIGH STRENGTH STEELS

E.A. Steigerwald

1. INTRODUCTION

The failure of high strength steels when exposed to a critical combination of stress and environment often represents a limiting design problem. In analyzing the nature of environmentally-induced failure the basic fracture resistance of the material must be considered along with its behavior in the degrading environment. The purpose of this presentation is to briefly review the characteristics of environmentally-induced fracture in high strength steels. Emphasis will be placed on considering the susceptibility to degrading environments when a precrack is present in the steel specimens. This approach allows the effects of specimen geometry to be minimized and provides the greatest sensitivity to both the external test conditions and the inherent material variables. The data should provide necessary background for defining the characteristics of the stress corrosion behavior and for evaluating the effectiveness of various methods for minimizing the delayed failure problem.

2. NATURE OF THE FAILURE PROCESS

Environmentally-induced failures occur by a process of crack initiation followed by slow, stable crack extension until the crack reaches a size which is critical with respect to the applied load. At this point rapid fracture occurs. The degree of slow crack extension which a material can tolerate is a direct function of its fracture toughness, the tougher materials require a more extended crack growth period. The initiation phase of fracture is generally attributed to localized pitting which develops into a crack-like defect. The pitting stage of fracture, which is the most time consuming phase of environmental cracking, usually requires 3 to 5 orders of magnitude more time than the stable growth period[1]. This characteristic makes the use of smooth test specimens often an inefficient method for studying environmental cracking. An example of the difference in failure time obtained from smooth and precracked 4340 steel* specimens is shown in Table I. When evaluating the relative susceptibility of a particular metal-environment couple the use of a precrack has the advantages of providing a sensitive short-time test with a minimum of scatter. As a result of these characteristics the precracked specimen is being widely used in environmental cracking studies[2,3,4].

In addition to providing a technique that provides meaningful results in relatively short times, the use of a precracked specimen along with linear elastic fracture mechanics analyses allows specimen geometry effects to be properly explained and produces definitive crack growth rates. A typical delayed failure curve obtained in precracked sheet specimens of 4340 steel at a 240 ksi tensile strength level exposed to a distilled water environment is shown in Figure 1. Delayed failure occurs over an extended region of applied stress and a well defined stress value exists below which failures are not observed in times of engineering significance. This lower critical limit when expressed in terms of the applied stress intensity level (K_{Iscc}) is independent of specimen geometry[5,6]. Although the use of K_{Iscc} has been employed as a design parameter[7] it is often considered too restrictive and some type of protective method is normally employed. In tests conducted on precracked

* The compositions of the specific steels referenced in this report are given in Appendix I.

specimens, a relatively short incubation time usually exists prior to the initiation of slow crack growth. Although still a matter of controversy this incubation period has been associated with the time required for the environment to reach the crack tip or a critical distance in the metal beneath the tip[2,8].

The distinct difference between the surface appearance of the stable environmental crack and the fast fracture is shown in Figure 2 for precracked bend specimens and Figure 3 for smooth specimens loaded in tension. The environment cracking usually has a very irregular surface topography, possibly as a result of the tendency of the crack to undergo various degrees of branching (see Figure 4). In certain high strength steels such as 4330V, 9Ni-4Co-30C, and 9Ni-4Co-45C (bainitic) exposed to a 3.5% NaCl solution, extensive macroscopic branching has been observed such that the main crack consisted of several paths[9]. The behavior has been attributed to a preference for the environmental cracking to follow the boundaries of the plastic zone.

The advantages of precracked specimens as a tool for studying the characteristics of environmental cracking have been previously reviewed in detail[1,10]. The importance of precracked specimens can generally be summarized as:

1. The time required to initiate environmentally-induced cracking is longer than the time for failure by crack propagation by a factor of about 10^5. This means that the majority of time in a stress-corrosion test is spent for the initial development of the stress-raiser and an overwhelming bias toward crack initiation is given in the time-to-rupture parameter In addition, the formation of a pit is often a statistically-controlled event which results in excessive scatter. The use of a precrack reduces the scatter and simplifies the analysis of a mechanism.

2. A given combination of environment and alloy may not produce pitting and may therefore indicate immunity in a test of a smooth specimen. If a stress-raiser is present, however, environmentally-induced failure may occur.

3. The presence of the precrack introduces the sharpest possible stress-raiser, hence from an analysis and design standpoint it results in conservative predictions.

4. The precrack provides a stress-raiser which is amenable to treatment by the linear elastic fracture mechanics equations and allows crack growth kinetics to be accurately evaluated.

The analysis of failure in precracked specimens has shown that the environmental embrittlement in high strength steels exposed to aqueous solutions is extremely localized at the crack tip and the properties of the remaining material are unaffected[11]. On this basis, the transition from slow to catastrophic crack growth is governed by the same fracture toughness which is measured without the embrittling action of the environment. Results shown in Figure 5 for high strength steel sheet specimens in gaseous atmospheres having varying water vapor contents further supports this characteristic by indicating that the threshold stress intensity (K_{Iscc}) (lower curve) increases with decreasing humidity but the stress intensity at which final fracture occurs is a constant, characteristic of the material and independent of environment.

In tests where extensive branching occurs or where the environmentally-induced crack front is extremely irregular (see Figure 6) the fracture mechanics equations which are derived on the basis of a single uniform check do not accurately describe the stress intensity along the crack tip. In these cases the apparent fracture toughness of the material is greater than that measured with a plane crack front and the environmental cracking will occur at a slower average rate.

In constant load tension specimens, the stress intensity at the crack tip increases as the crack extends. Since the crack growth kinetics are controlled by the stress intensity, a test of this type results in a constantly increasing crack growth rate as a function of test time. In certain cases the analysis of crack growth kinetics is simplified by using

a specimen geometry which is designed so that the stress intensity value (K_I) at the crack tip is constant as a function of crack length when a constant load is used. In materials where the crack does not have a tendency to move out of the center plane of the specimen, this specimen type is capable of yielding a maximum amount of data on crack growth rates and threshold stress intensity values[13].

Although the precrack specimens have a large number of advantages certain disadvantages also exist particularly when methods for minimizing stress corrosion effects, such as coatings, are being evaluated. The presence of a relatively long precrack makes uniform coating application often difficult. In cases where superimposed electric potential is applied the quantitative value of polarity and solution pH is also difficult to meaningfully define.

3. EFFECT OF EXTERNAL VARIABLES

The external variables of applied stress, environment, electric potential, and temperature, all exert a significant influence on the environmental cracking problem. The general relationship between stress and delayed failure was shown in Figure 1. One significant point is that the dependence of the environmentally-induced failure process is relatively insensitive to stress over the failure range by comparison with other time dependent process such as creep or fatigue. A second factor is that a well-defined lower critical limit exists (K_{Iscc}). In a delayed failure test on a precracked specimen the time to failure is controlled by the rate of slow crack growth and the total distance over which the crack must grow to produce failure. As shown in Figure 7 the crack growth rate is virtually independent of stress in high strength steels over a rather extensive range of applied stress values. As a result the slight variation in failure time is almost a direct function of the fact that the crack must grow a slightly longer distance at the lower applied stress to produce fracture. Although there is no accepted explanation for the insensitivity of the slow crack growth rate to the applied stress, the behavior is similar to that observed for delayed failure of hydrogenated high strength steels[14].

A number of studies have been conducted on the influence of environment on delayed failure in high strength steels. The results can be generally summarized as follows:

1. Distilled water and aqueous chloride solutions give comparable delayed failure behavior[15].

2. Water vapor in an inert gaseous environment can produce environmental cracking[2,3].

3. Dry hydrogen gas at one atmosphere pressure will produce delayed failure in precracked specimens at a rate which is faster than a moist air environment. The crack growth rate in dry hydrogen can be retarded by adding small quantities of oxygen to the environment[16].

4. Variations in solution pH do not produce a marked effect on the environmental crack growth rate[13].

5. High strength steels respond differently to applied electric potential. In some material-environment systems cathodic protection is observed over a limited range and anodic acceleration occurs, while in other systems only cathodic acceleration of the crack growth rate takes place[15].

6. Delayed failure has been observed in organic environments and the failure time increases as the solubility of water in the environment increases[17].

The behavior of 4340 steel in distilled water and 3.0N NaCl solution, presented in Figure 8, indicates that both environments are equally effective in producing environmental cracking. Although there have been instances where differences have been observed between chlorides and distilled water no consistent pattern has been established. The

influence of water vapor content on the delayed failure of precracked steel specimens in an air environment is shown in Figure 9. Above approximately 60% relative humidity the specimens behaved as if they were in a liquid environment, presumably because water actually condenses at the crack tip. The ability of dry hydrogen gas at one atmosphere and humidified argon to produce slow crack growth has been demonstrated by the classical work of Hancock and Johnson[16] (see Figure 10). The crack growth rate could be retarded by adding a small quantity of oxygen to the environment.

The influence of variations in solution pH on the crack growth rate of 4340 (260 ksi tensile) is shown in Figure 11 with applied potential plotted parametrically. A significant effect of pH was only noted at a large cathodic potential or at a high pH value when relatively high stresses were applied to the specimen.

In many instances the application of an electric potential has been used in an effort to define the environmental cracking mechanism or to examine the possibility of using cathodic protection as a method for inhibiting cracking. The delayed failure characteristics of 4340 and D6AC steel as a function of applied potential are shown in Figures 12 and 13. The behavior pattern of precracked high strength steel specimens exposed to an electric potential is not consistent. In fact, accurate interpretations of the significance of the potential in defining a mechanism are complicated by the fact that conditions at the tip of a crack may be considerably different than the gross parameters measured on the bulk of the specimen. This point will be further discussed in the section dealing with possible mechanisms for environmental cracking.

A summary of the influence of test temperature on environmental cracking in high strength steels is given in Figure 14 and Table II. For water environments the crack growth rate generally increases with increasing temperature while the reverse is true for gaseous environments[2]. Although differences in viscosity and surface behavior have been suggested to accound for the variation between the two environments as a function of temperature, no accepted explanation has been developed. The results summarized in Table II indicate the similarity between the activation energy for hydrogen diffusion, hydrogen embrittlement, and environmental cracking in high strength steels.

4. EFFECT OF MATERIAL VARIABLES

The effect of composition on the environmental cracking susceptibility of high strength steels is often difficult to evaluate since the compositional influence is usually confounded by concomitant changes in strength level and structure. A series of delayed failure curves for several martensitic high strength steels heat treated to a constant tensile strength of 235 ksi, exposed to a distilled water environment and tested in the precracked condition is shown in Figure 15. Although the steels experienced environmental cracking over a comparable stress range, the kinetics of the failure process were significantly altered by material composition. In general, there was almost a three order of magnitude difference between the relatively low alloyed 4340 and the more highly alloyed steel. Despite this change in kinetics the susceptibility to delayed failure in terms of stress range was not significantly altered. The composition effect is further illustrated in Figure 16 by comparing the incubation time prior to the start of crack growth and the failure time at approximately 60% of the fracture stress intensity (K_c). These values are plotted against the arbitrarily selected correlating parameter 5Mo + 2.5Cr + Ni + Co*. The approximate linear relationship between $\log t_f$ (or t_i) and the compositional parameter suggests that there is a systematic influence of compositional factors on the kinetics of environmental cracking in the specific martensitic steels tempered to a structure that produces a given strength level.

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* Alloy contents are expressed in weight percent.

The more highly alloyed maraging steels generally exhibit improved resistance to environmental cracking. The variation of K_{Iscc} with strength level is shown in Figure 17 for a number of high strength steels. The maraging grades are usually superior to the lower alloy steels that depend on carbon for strength.

In addition to compositional effects, metallurgical structure can significantly influence the susceptibility of a steel to delayed failure. The effect of structure has been studied primarily in the 9Ni-4Co steels, where variations in carbon level and/or heat treatment can produce several different structures with comparable strength levels. Failure and incubation times for martensitic HP 9-4-25, bainitic HP 9-4-45 and martensitic HP 9-4-45 steels at tensile strength levels of approximately 235 ksi are plotted as a function of the K_i/K_c ratio in Figure 18. Failure times and incubation times for the bainitic 9-4-45 and 9-4-25 were comparable at all K_i/K_c ratios. Below a K_i/K_c ratio of 0.7, the incubation times for the martensitic 9-4-45 were significantly shorter than the low carbon or bainitic structure while failure times were shorter for the martensitic 9-4-45 alloy at all K_i/K_c ratios.

If composition alone is the primary cause of the difference in the kinetics of delayed failure, the various structures in the 9Ni-4Co steels would exhibit comparable stress corrosion behavior. The appreciable difference in failure kinetics produced by heat treatment or carbon content indicates the significant effect that can be exerted by structural control.

Martensitic HP 9-4-45 is reported to have a twinned martensite structure, while bainitic HP 9-4-45 is not twinned[19]. Although the specific structure of HP 9-4-25 has not been examined, information available on the HP 9-4-20 and HP 9-4-30 alloys suggests that HP 9-4-25 does not have a twinned structure. Furthermore, in the 9-4-45 martensitic alloy, the twinned martensite is reported to have a heavy concentration of carbides at platelet boundaries as well as within the platelets. Untwinned bainite at the same strength level has a more uniform distribution of carbides and no indicated preference for precipitation at platelet boundaries[19]. There is no information available on carbide morphology in the 9-4-25 alloy.

These differences are sufficient to suggest that the delayed failure kinetics in the 9Ni-4Co steels are regulated by martensite and/or carbide morphology. The twinned structure or the preferential carbide distribution in the 9-4-45 martensite provide a network of high energy boundaries which could produce a preferred path for crack propagation, thus accelerating the crack growth kinetics.

5. POSSIBLE FAILURE MECHANISMS

At present, environmentally-induced failure of metals under sustained load can be attributed to one or more of the following three mechanisms:

1. Anodically-induced stress corrosion.
2. Hydrogen embrittlement.
3. Stress sorption cracking (lowering of the surface energy in the crack due to liquid adsorption).

The most widely known theory, the electrochemical theory, was developed by Dix[20]. He proposed that selective corrosion of a solute depleted matrix, together with a high stress acting to pull the metal apart, are necessary conditions for stress corrosion cracking. The stress is required to destroy protective films and expose fresh anodic materials to the corrosive medium. The condition which defines electrochemical stress corrosion as the responsible mechanisms is usually whether or not the application of a negative potential retards the embrittlement. Although it is generally agreed that a mechanism of this type

defines the behavior present in austenitic stainless steels, controversy still exists concerning the role of hydrogen in the failure process.

A mechanical theory of stress corrosion cracking recently proposed by Nielson[21] is based upon the fact that corrosion products occupy a larger volume than the volume of metal destroyed. They would, therefore, introduce a simple wedging action which, in time, would be sufficiently large to cause crack propagation. The process is then repeated several times before failure occurs. This mechanism would suggest discontinuous crack propagation, as well as an incubation time before slow crack growth is initiated.

Cracking of steels in the presence of corrosive media has also been attributed to embrittlement by hydrogen, which is supplied by the corrosion reaction or by cathodic polarization. Hydrogen embrittlement tends to produce cracking which may be either transgranular in nature, or follow former austenite grain boundaries. The characteristics of hydrogen cracking include: (a) a specific incubation time before the initiation of crack propagation, (b) a minimum stress below which delayed failure will not take place, (c) discontinuous crack propagation, (d) an acceleration of the embrittlement by cathodic polarization, and (e) a reversibility of the incubation time with respect to applied stress.

A series of tests involving reversibility, superimposed currents, and poisons have been conducted which indicate that the delayed failure of low alloy martensitic high strength steel in distilled water or moist air environments was due to a hydrogen embrittlement mechanism where the nascent hydrogen was formed by the corrosion reaction[2]. Further confirmation that a hydrogen embrittlement mechanism is operative in the high strength steels is obtained from the work which showed that delayed failure takes place in precracked specimens exposed to a dry hydrogen environment at room temperature[16].

A mechanism called stress sorption cracking, is frequently used, at least in part, to explain embrittlements due to corrosive environments. In principle, this theory involves adsorption of an atom or molecule, which reduces the surface or bonding energy of the metal at the apex of the crack and allows the metal to fail under the applied tension forces[22]. In cases where the adsorbing atom is hydrogen, the stress sorption theory is just a variation of hydrogen embrittlement.

At present the hydrogen embrittlement theory has received the greatest degree of support as the mechanism for environmental cracking of high strength steels exposed to aqueous environments. Although the data which show that anodic polarization of precracked steel specimens can accelerate cracking have been interpreted as incompatible with hydrogen embrittlement, recent work has shown that hydrogen can be generated locally at the base of pits or cracks even under anodic conditions[23]. The variation of hydrogen permeability with polarization is shown in Figure 19 superimposed on the environmental cracking susceptibility. The correlation between hydrogen generation (as measured by permeability tests) and delayed failure in the anodic region removes a series obstacle to the application of the hydrogen embrittlement mechanism. The following observations are believed to provide adequate support that hydrogen embrittlement is the critical mechanism for the environmentally-induced delayed failure of high strength steels.

1. The qualitative delayed failure of hydrogenated specimens is directly comparable to failure in aqueous environments, e.g., crack growth rates, discontinuous cracking, and a lower critical limit.

2. The activation energy for crack growth is comparable for hydrogenated steel specimens and specimens exposed to distilled water. This activation energy is not altered by superposition of anodic or cathodic potential.

3. Hydrogen permeability can be directly correlated with environmental cracking over a wide range of polarization values.

4. The inclusion of additions to the environment which are known to facilitate the absorbtion of hydrogen also increases the environmental cracking susceptibility.

5. Dry hydrogen gas at one atmosphere is an extremely degrading environment to pre-cracked steel specimens.

6. SUMMARY AND CONCLUSIONS

The delayed failure of high strength steel when exposed to a static load and an aqueous environment often represents a critical design consideration. The failure is characterized by three distinct parameters: (1) the strength which represents the limit of the load carrying ability of the specimen in the absence of environmental effects, (2) the lower critical stress below which failure is not observed in times of engineering significance, and (3) a region of crack initiation and slow crack extension which leads to total specimen failure when the critical combination of applied load and crack length is attained.

Studies have been conducted to determine the influence of external variables such as specimen type, environment, and temperature; and material variables such as composition and structure on the characteristic behavior of the delayed failure process in high strength steels. Particular emphasis has been placed on using the applied stress intensity parameter obtained from linear elastic fracture mechanics analyses to separate the environmentally-induced slow crack growth from the limitations produced by the inherent material toughness. The stress intensity parameter at the lower critical limit (K_{Iscc}) is independent of specimen type. Because of the higher critical stress intensity for failure in the absence of an environment, sheet specimens develop a more extended delayed failure region. The influence of steel composition on the delayed failure characteristics have been evaluated for a variety of quenched and tempered martensitic high strength steels. The alloy content of the martensitic steels strongly influences the kinetics of the delayed failure process but has only a limited effect on the range of susceptibility. The role of structure in the failure process has been evaluated with the 9Ni-4Co steels. Bainitic heat treatment of 9Ni-4Co-45C or quench and temper treatment of 9Ni-4Co-25C result in a significant improvement in delayed failure resistance relative to the quenched and tempered 9Ni-4Co-45C. The reported absence of a twinned structure in the 9-4-25 alloy and the 9-4-45 bainite has been suggested as a possible reason for the improved resistance to environmental failure. The maraging steels which do not depend on carbon content to develop the desired strength have significantly greater delayed failure resistance than the quenched and tempered martensitic steels.

Experiments which involved studying the crack growth behavior, the activation energy, and the reversibility of the incubation time under anodic and cathodic polarization indicate that the characteristics of each of these parameters under cathodic conditions, where hydrogen embrittlement is the accepted mechanism is directly comparable to that obtained under anodic polarization. Hydrogen permeability tests, performed over the same range of polarization as the delayed failure tests, showed that hydrogen could be generated locally under anodic conditions and a direct correlation existed between the tendency for delayed failure and the hydrogen permeation. The overall results of the studies are consistent with a failure mechanism which involves embrittlement of the steel by the hydrogen generated in the corrosion process.

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TABLE I

Comparison of Environmental Cracking Test Duration Using Bent Beam Specimen Without a Precrack and Precracked Tensile Specimens, 4340 Steel, Distilled Water Environment

Specimen Type	*Material Yield Strength (ksi)*	*Applied Stress (ksi)*	*Failure Time (hours)*
Bent Beam	220	176	1000
Bent Beam	220	176	No failure in 4500
Bent Beam	225	180	No failure in 2000
Bent Beam	240	192	700
Center Precracked Tensile	213	40	0.17
Center Precracked Tensile	213	40	0.13
Center Precracked Tensile	208	40	0.08
Center Precracked Tensile	185	60	82

TABLE II

Activation Energies for Hydrogen Diffusion and Environmental Cracking in High Strength Steel[12]

Investigator	*Experimental Description*	*Activation Energy*
Beck, Bockris, McBreen, Nanis[18]	Diffusion of electrolytic hydrogen through AISI 4340 membrane	9220 cal/g-atom
Steigerwald, Schaller Troiano[14]	Incubation period for crack initiation in cathodically charged notched rounds	9120 cal/g-atom
Johnson and Willner[3]	Center-cracked plates in water and saturated water vapor (H-11)	9000 cal/g-atom
Van der Sluys[13]	Precracked double cantilever beam in water (4340)	8500 cal/g-atom

APPENDIX I

Nominal Chemical Compositions of Steels Described in this Presentation (Wght.%)

Steel	*C*	*Mn*	*Si*	*Ni*	*Cr*	*Mo*	*V*	*P*	*S*	*Fe*	*Co*
H-11	0.42	0.29	0.96	–	5.17	1.35	0.52	0.021	0.006	Bal.	
4340	0.40	–	–	1.85	0.90	0.40	–	–	–	Bal.	
300M	0.40	–	1.75	1.85	0.90	0.40	–	–	–	Bal.	
18% Ni Maraging (18-7-5)	0.007	0.03	0.01	18.0	–	4.82	–	–	–	Bal.	7.45 0.39 Ti, 0.10 Al
HP 9-4-45	0.43	0.15	0.01	7.80	0.32	0.30	0.09	0.006	0.009	Bal.	3.95
HP 9-4-25	0.25	0.15	0.01	7.80	0.32	0.30	0.09	0.006	0.009	Bal.	3.95
D6AC	0.45	0.69	0.26	0.55	1.0	1.0	0.08	0.008	0.006	Bal.	
4330V	0.30	–	–	1.85	0.90	0.40	0.52	–	–	Bal.	

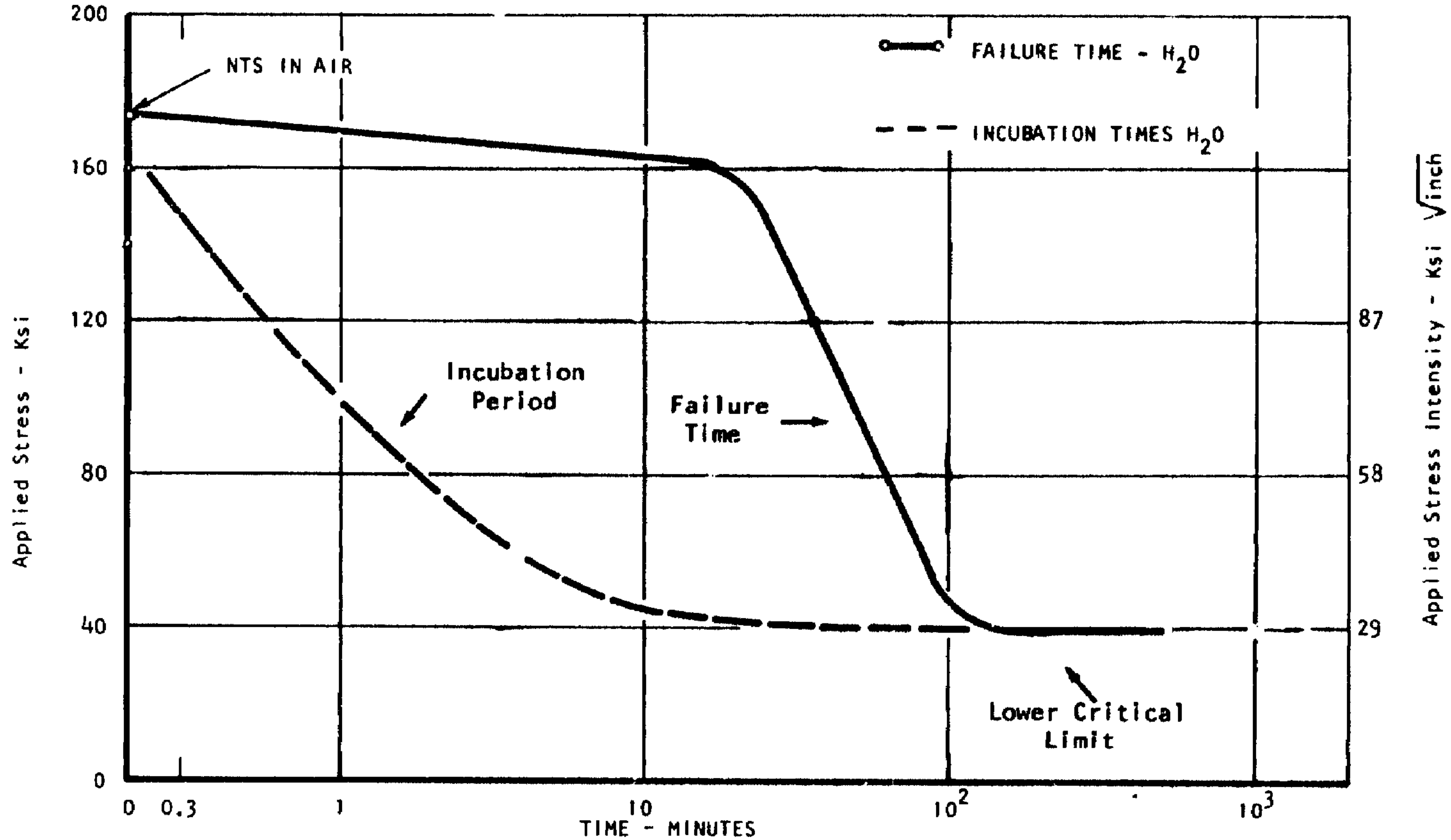
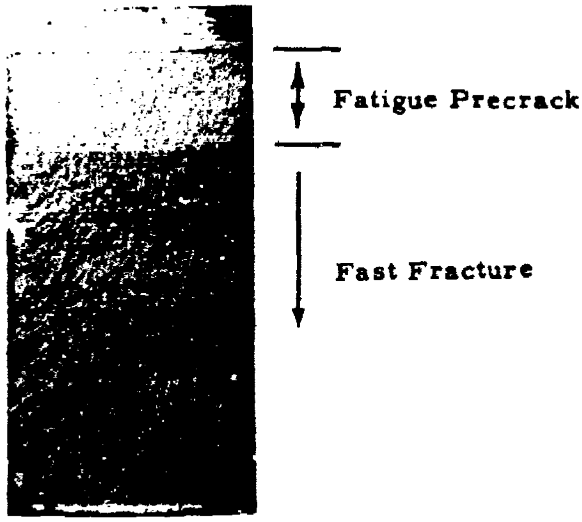
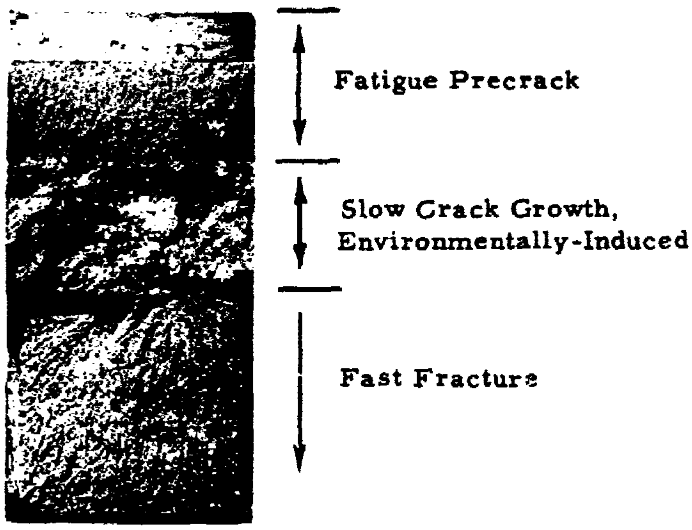


Fig.1 Delayed failure of AISI 4340 steel (240 ksi strength level) precracked sheet specimens in distilled water



Air Environment
$K_{IC} = 73.3$ Ksi $\sqrt{in.}$ (continuously increasing load)



Distilled Water
Environment
$K_i = 62.2$ Ksi $\sqrt{in.}$ (statically loaded)

Fig. 2 Fracture appearance of 4340 cantilever beam precracked bend specimens (235 ksi strength level)

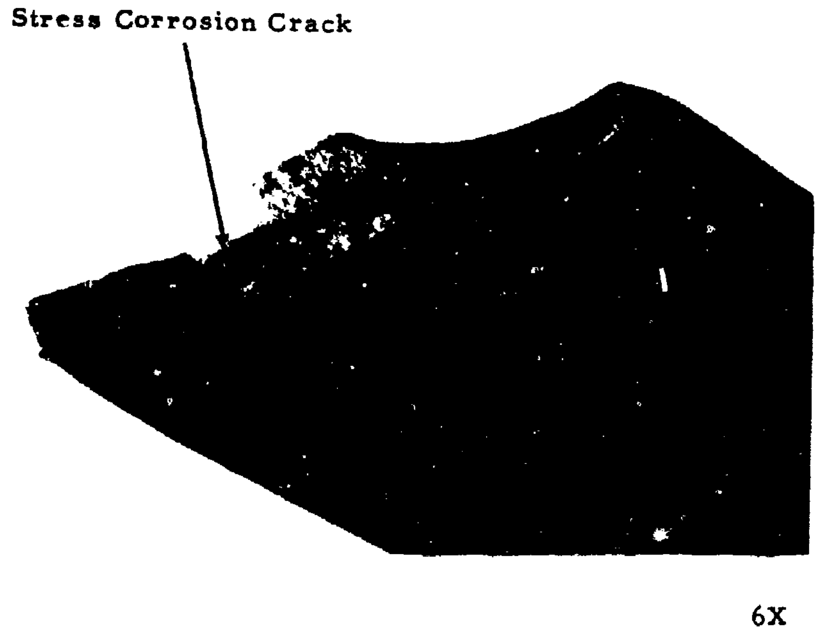


Fig. 3 Fracture appearance of 4340 (235 ksi strength level) smooth specimen which failed after 510 hours at an applied stress of 0.96 of the yield strength, distilled water environment, room temperature test

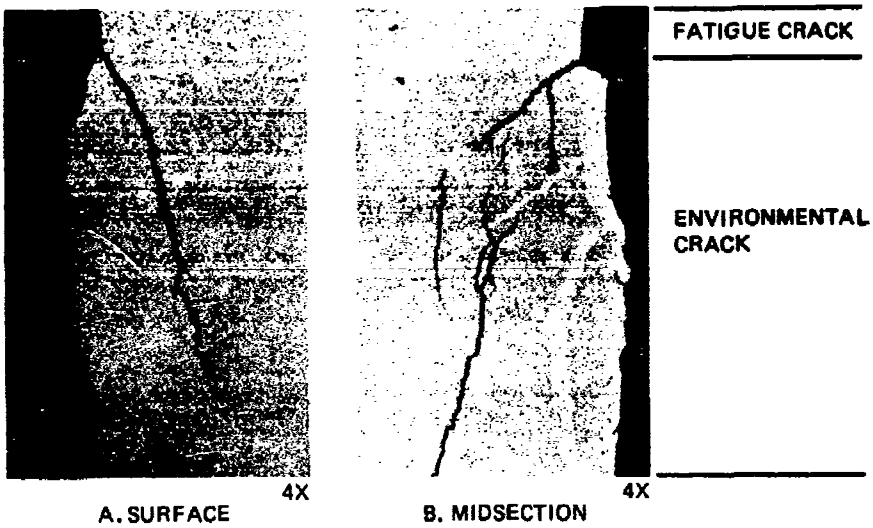


Figure 4A. *Section of failed 4330V notched bend specimen (K_{Ii} = 83.9 ksi$\sqrt{in.}$)*

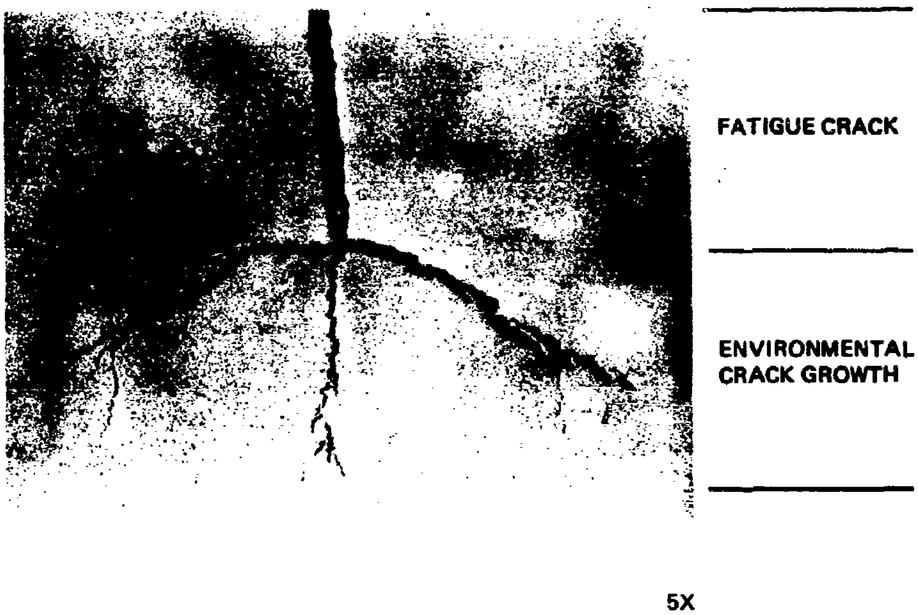


Figure 4B. *Unfailed 4330V notched bend specimen (K_{Ii} = 99.0 ksi$\sqrt{in.}$)*

Fig. 4 Examples of crack branching in high strength steel exposed to a 3.5% sodium chloride environment (Ref. 9)

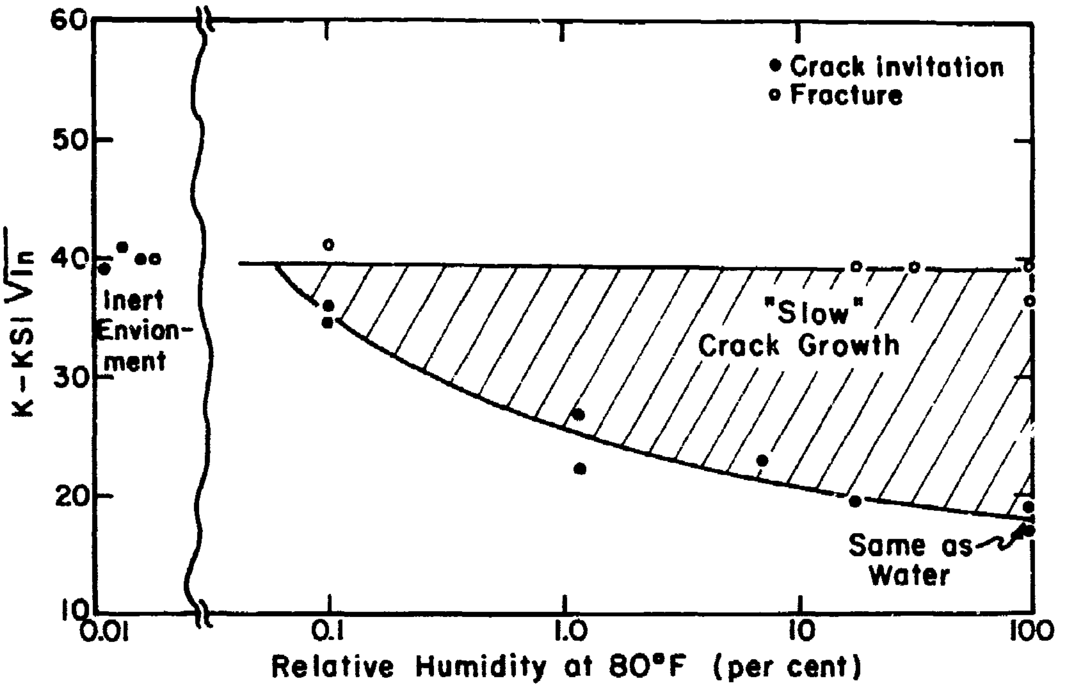


Fig. 5 Variation of crack growth initiation and fracture parameters with relative humidity (Ref. 12)

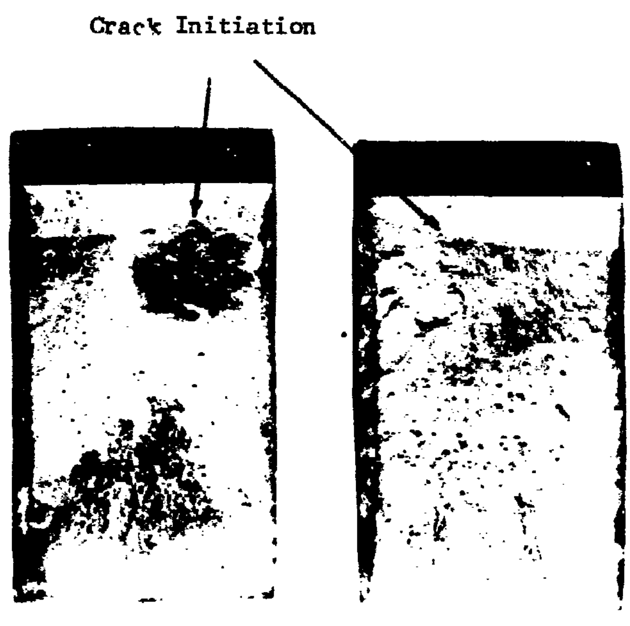


Fig. 6 Examples of irregular fracture surfaces of HP 9-4-45 steel, precracked bend specimens

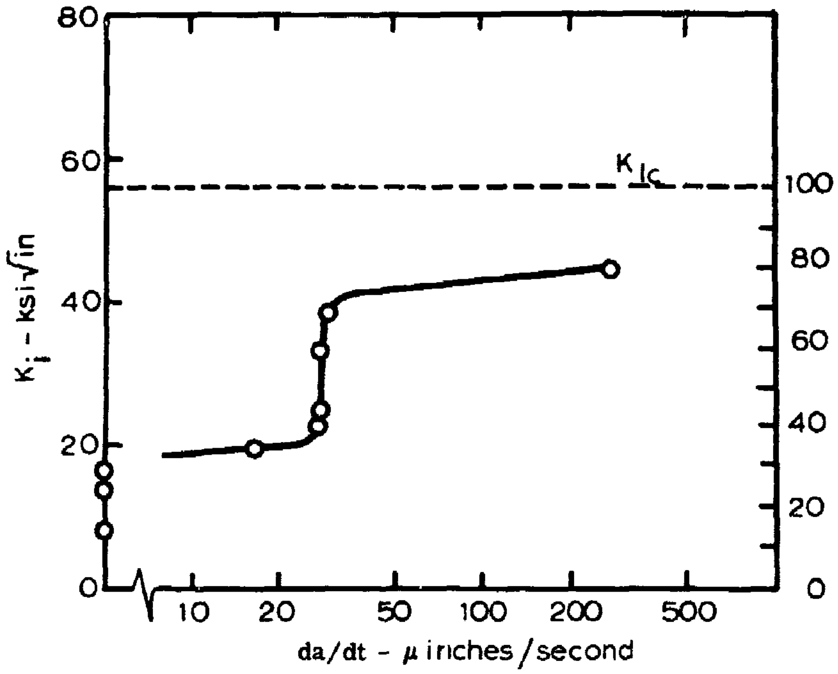


Fig. 7 Crack growth rate, da/dt , versus applied stress intensity factor, K_i , and ratio of K_i to K_{Ic} for AISI 4340 M steel, tempered for 2 hours at 550°F. Specimen immersed in a 3-1/2% NaCl solution at room temperature (Ref. 4)

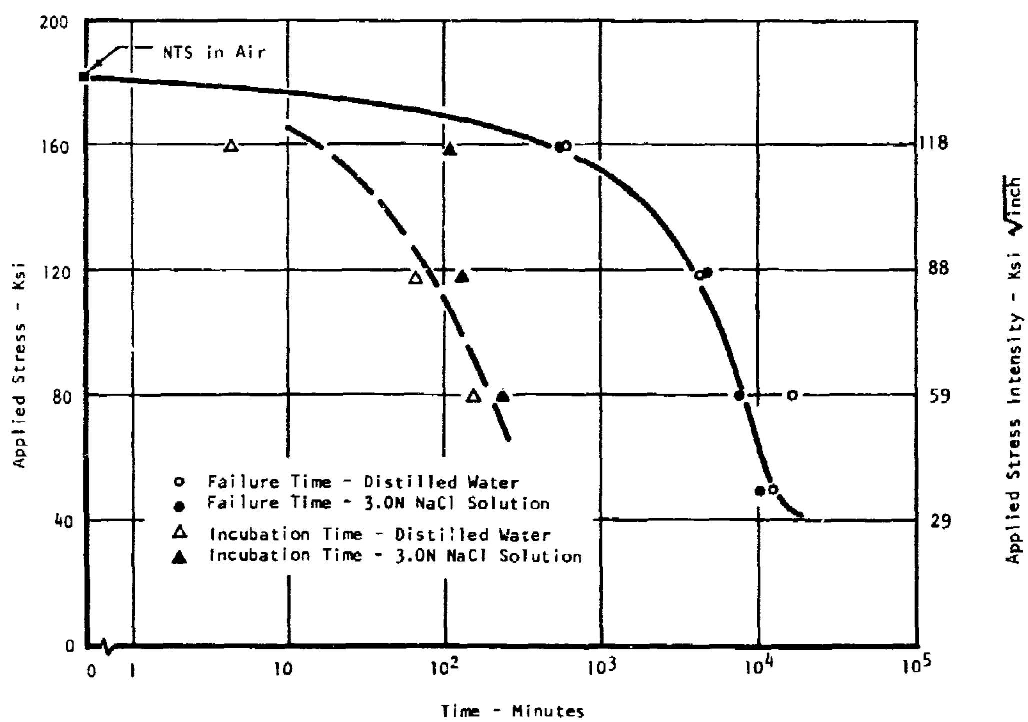


Fig. 8 Delayed failure of martensitic HP 9-4-45 steel (240 ksi strength level) in distilled water and 3.0N NaCl solution

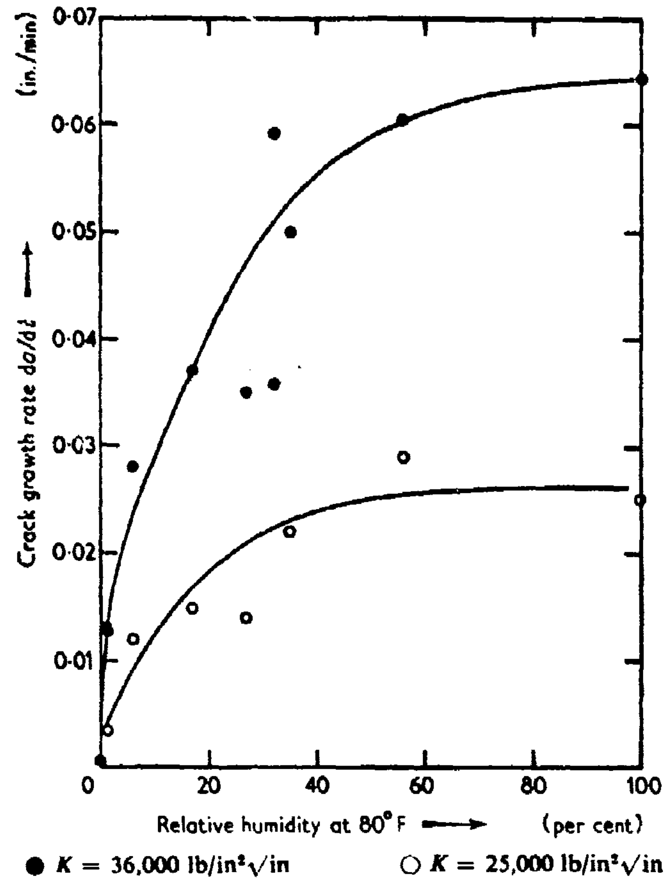


Fig.9 Crack growth rate versus relative humidity, H-11 steel (Ref.3)

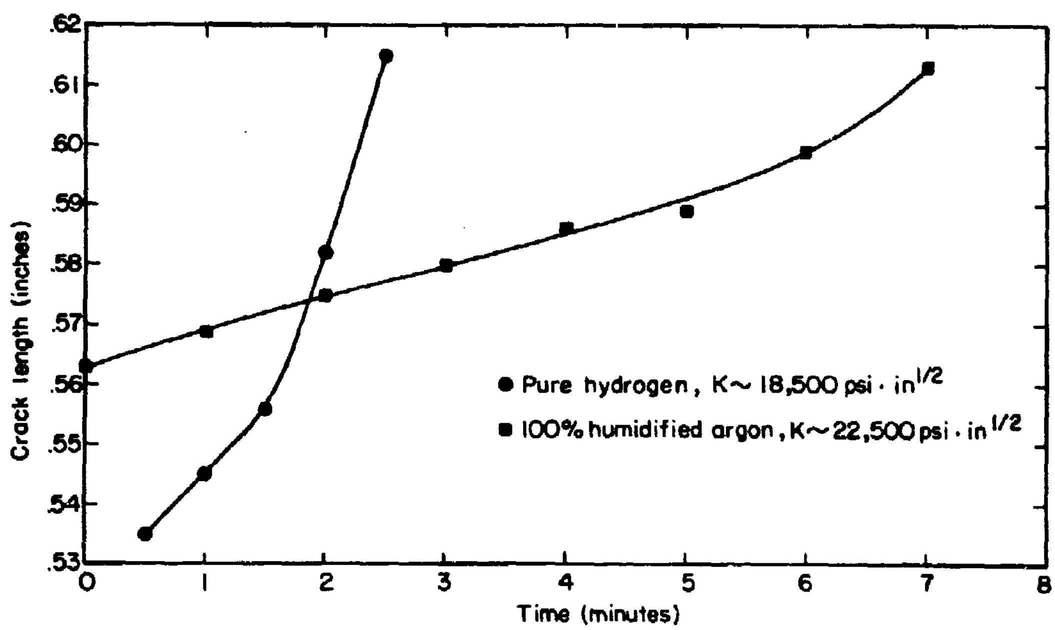


Fig.10 Subcritical crack-growth kinetics in hydrogen and humidified argon, H-11 steel (Ref.16)

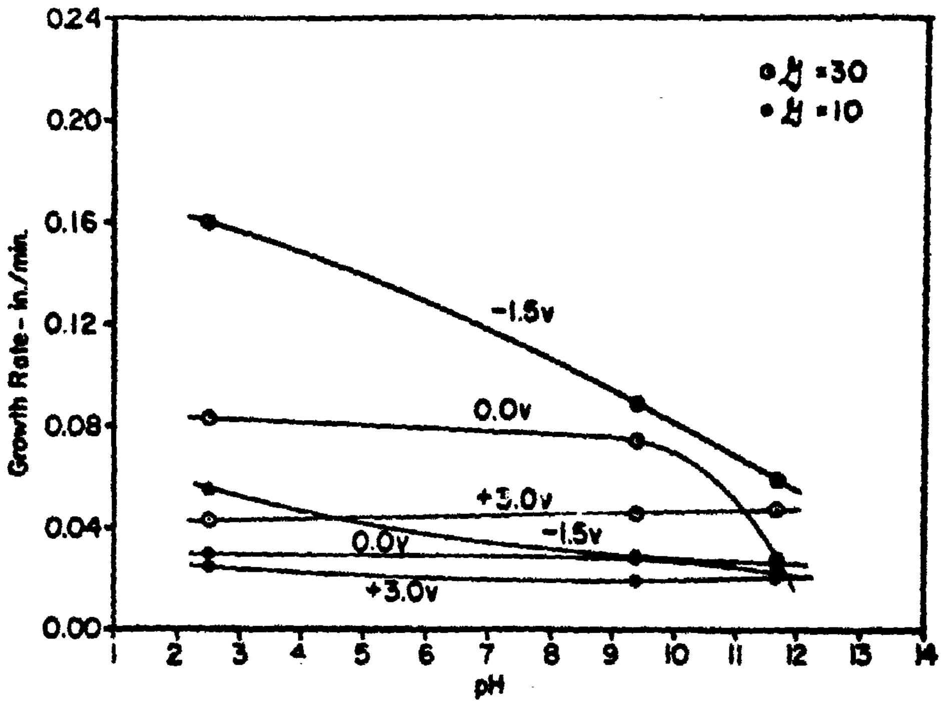


Fig. 11 pH versus crack growth rate, precracked 4340 steel specimens, water environment (Ref. 13)

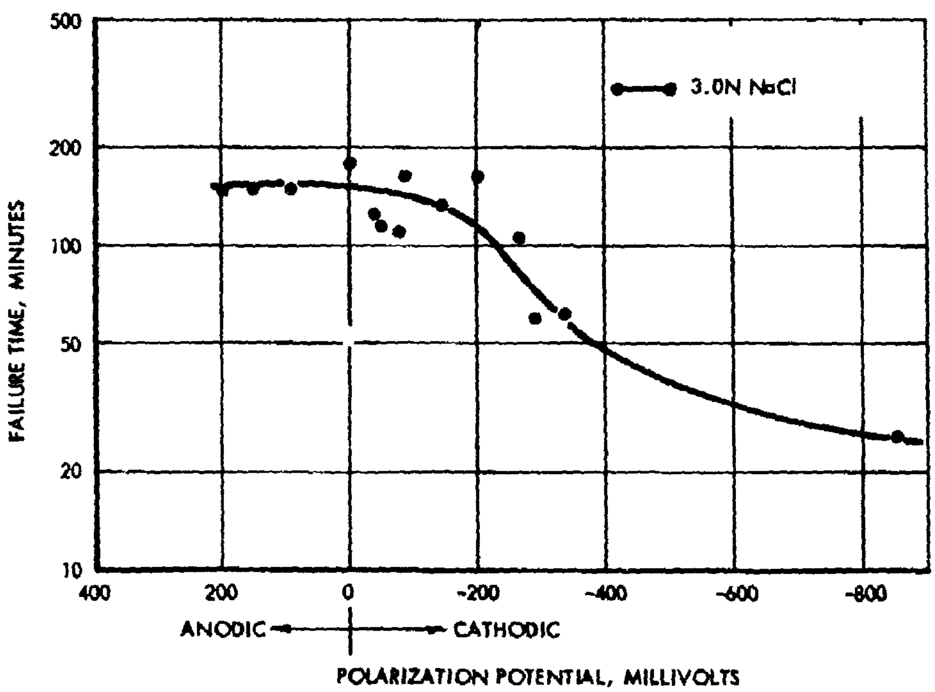


Fig. 12 Effect of impressed polarization potential on delayed failure of AISI 4340 steel (235 ksi strength level) center-notch specimens in 3.0N NaCl solution at 50 ksi applied stress

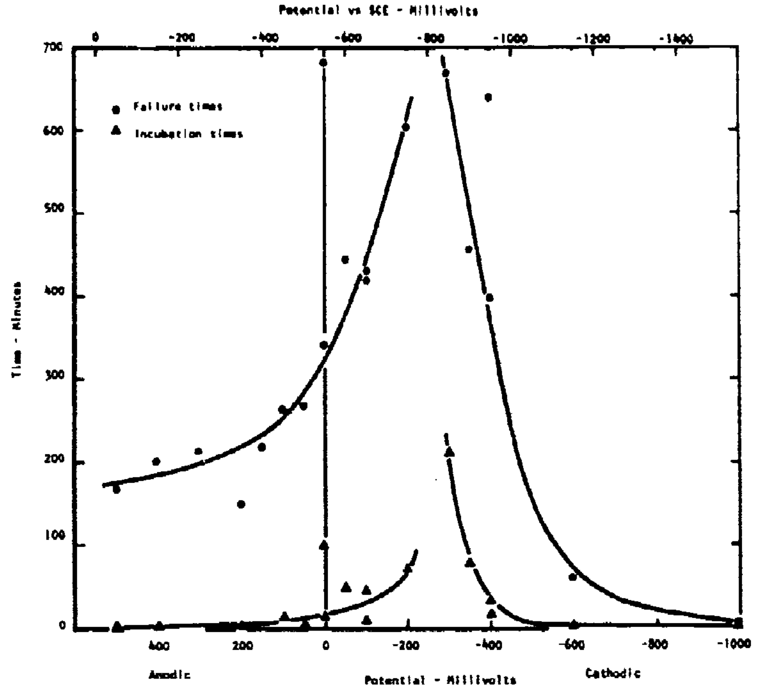


Fig. 13 Results of polarization experiments on D6AC steel, 120 ksi applied stress, 3.0N NaCl environments

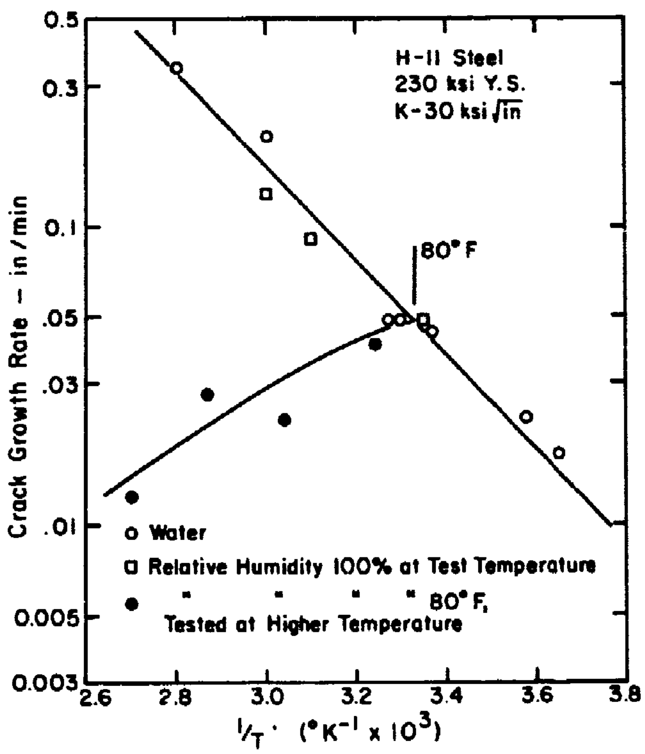


Fig. 14 Variation of crack growth rate with test temperature (Ref. 12)

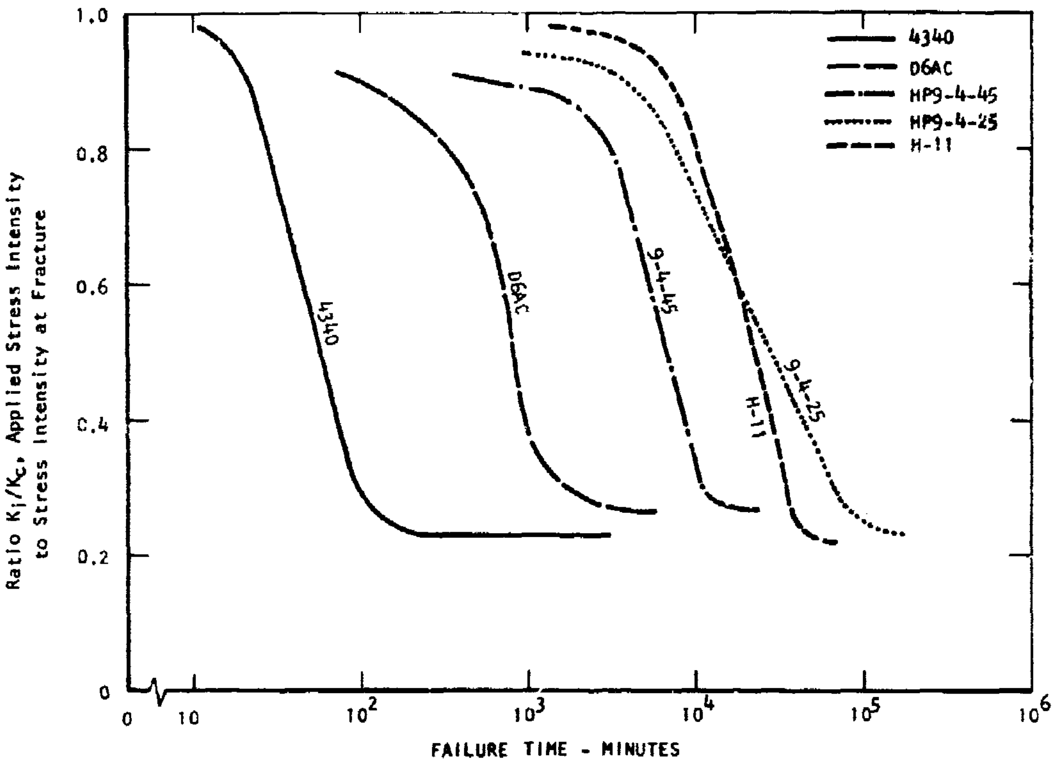


Fig. 15 K_i/K_c ratio versus failure time for five martensitic high-strength steels, distilled water environment

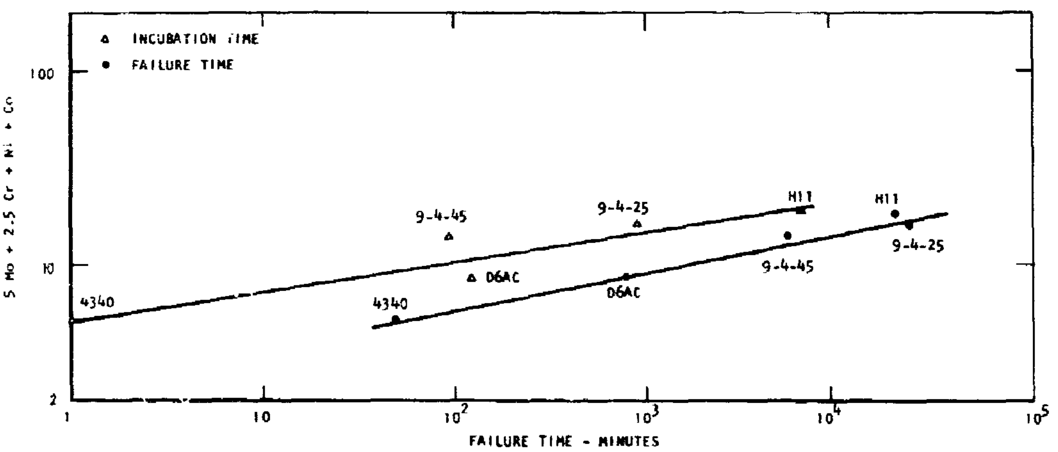


Fig. 16 Effect of composition on incubation time and failure time for five martensitic steels at $K_i/K_c = 0.60$ in distilled water

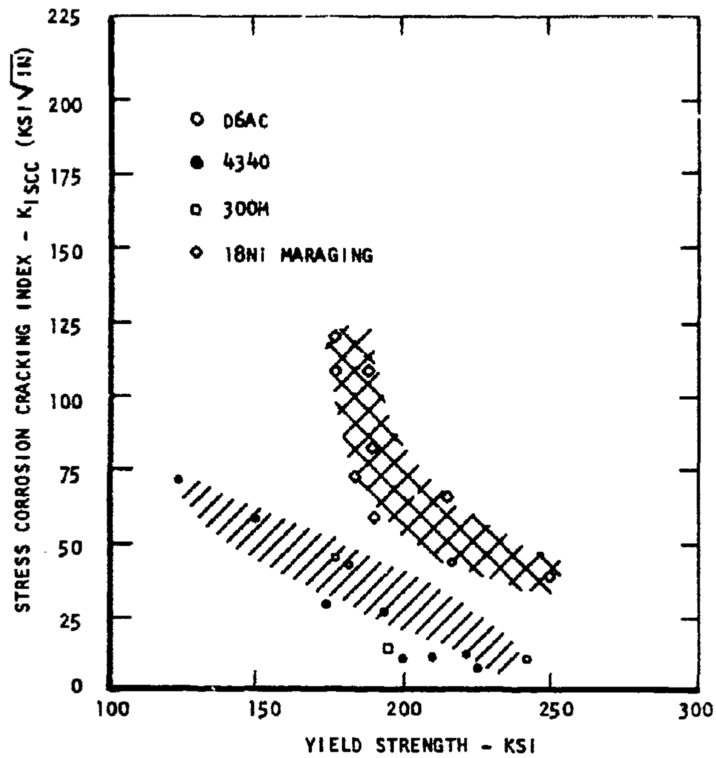


Fig.17 Variation of K_{Iscc} with yield strength for a variety of high strength steels (Ref.9)

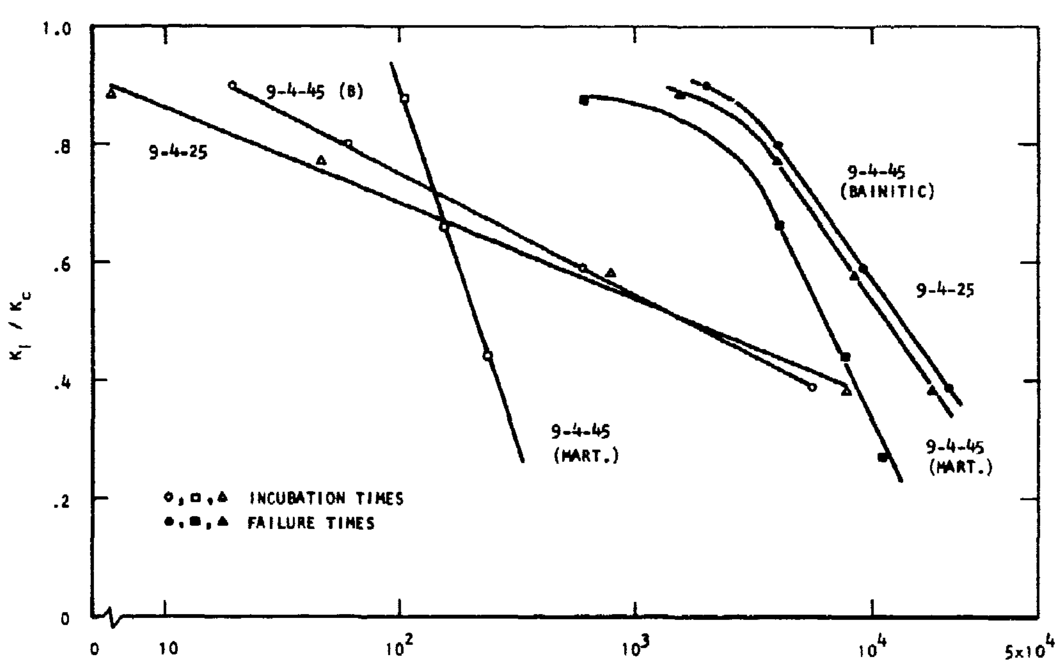


Fig.18 Comparison of delayed failure in HP 9-4 steels as a function of normalized applied stress intensity

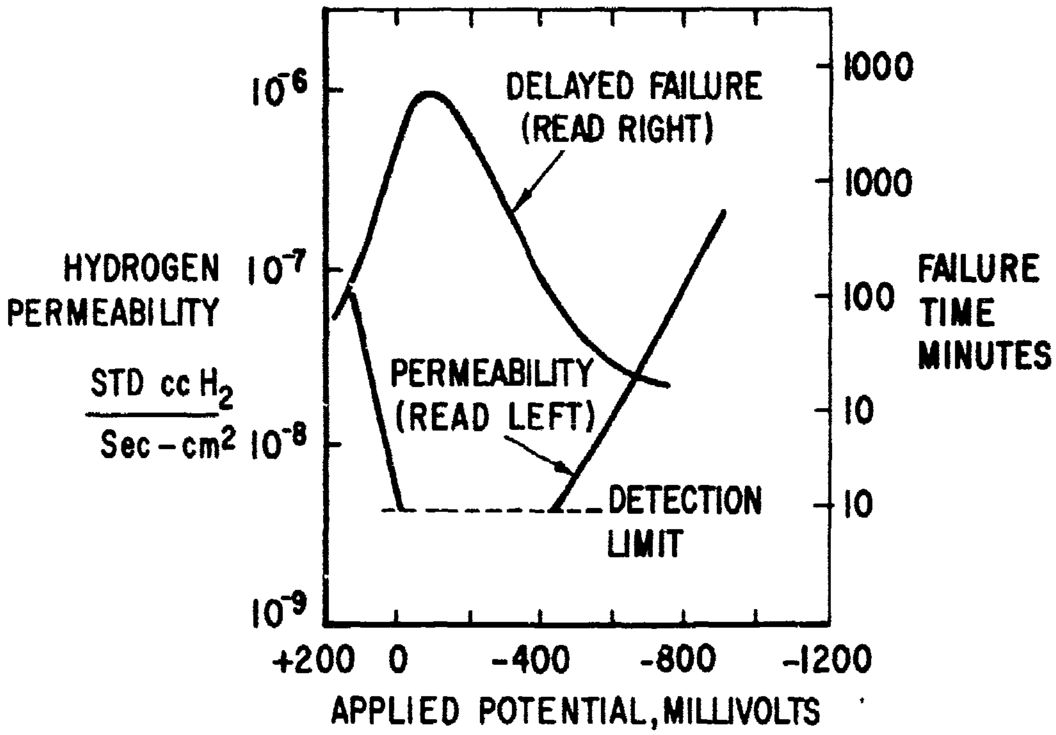


Fig. 19 Comparison of delayed failure data at an applied stress of 50 ksi and hydrogen permeability for 9-4-45 steel, precracked specimens

ETUDE DU COMPORTEMENT DE L'ALLIAGE MAGNESIUM - 8% d'ALUMINIUM EN CORROSION SOUS TENSION

par

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SOMMAIRE

A partir de cas de rupturesen service sur l'alliage de forge Magnésium - 8% d'aluminium, quelques essais ont été entrepris en laboratoire afin d'étudier le comportement de cet alliage en corrosion sous tension. Une comparaison a été faite avec un alliage de fonderie ayant sensiblement la même composition.

Contrairement aux alliages à base d'aluminium, l'alliage Mg Al 8% est sensible à la corrosion sous tension dans tous les sens de prélèvements et les fissures peuvent être soit intergranulaires, soit transgranulaires. Toutefois, la sensibilité de l'alliage est limitée, ce qui devrait permettre d'éviter assez facilement les ruptures en service.

Au nombre des problèmes soulevés par l'utilisation du magnésium et de ses alliages en construction aéronautique, la corrosion sous tension est rarement évoquée. On pense en général, avant tout, à protéger l'alliage contre la corrosion banale due aux agents atmosphériques, aux couplages avec d'autres métaux plus nobles, etc...

Toutefois, une pièce correctement protégée contre la corrosion et qui réussit, de ce fait, à demeurer un temps prolongé en service peut, dans certaines conditions, se rompre par corrosion sous tension. La corrosion du magnésium et de ses alliages a déjà fait l'objet d'études systématiques par Logan (1) (2) (3) et Fontana (4) aux Etats-Unis. Nous avons en France, à la Direction Technique des Constructions Aéronautiques, été à même d'examiner de nombreuses ruptures en service sur des pièces matricées en alliage GA8-Z1 (désignation ASTM AZ 80 A). Nous avons rassemblé dans cette communication les diverses observations que nous avons pu faire au cours de l'examen de ces pièces et une étude des conditions dans lesquelles on pouvait reproduire le phénomène en laboratoire. Les essais ont été effectués sur l'alliage de forge GA8-Z1 et sur un alliage coulé de composition voisine GA9.

1. RUPTURES EN SERVICE

La plupart des pièces examinées provenaient de commandes de vol d'hélicoptères utilisés par la Marine ou l'Armée de l'Air. Les premières ruptures se sont manifestées après 5 ou 6 ans d'utilisation, sur les pièces les plus diverses : bielles, secteurs de commande, chapes, guignols, etc... On en trouvera plusieurs exemples figures 1 à 5.

On remarquera que, dans tous les cas, les ruptures ou les fissures de corrosion sous tension n'intéressent que des pièces peu sollicitées en service, mais sur lesquelles des contraintes permanentes excessives ont été constatées :

- pincement anormal des chapes (fig. 1 et 2)
- contraintes de montage d'une bague ou d'un roulement (fig. 3 et 4)
- contraintes élastiques développées autour d'un poinçonnage (fig. 5)

Le relevé des déformations montrait très souvent que les contraintes de montage excédaient la limite élastique du matériau.

Dans presque tous les cas, les pièces avaient remarquablement résisté à la corrosion et les protections par peinture ou mordançage paraissaient intactes. Seules quelques criques s'amorçaient sur les piqûres de corrosion peu importantes (fig. 4).

2. ESSAIS de CORROSION SOUS TENSION en LABORATOIRE

2.1 Choix de la solution d'attaque

Pour reproduire les ruptures en laboratoire, il nous fallait travailler dans une solution qui provoque la corrosion sous tension sans entraîner une corrosion généralisée trop importante qui risquait de masquer le phénomène.

Des essais de corrosion ont été entrepris dans différents milieux :

- solution à 3% de Cl Na tamponnée au PH 8
- eau distillée
- eau du robinet

avec quelques-uns des inhibiteurs de corrosion proposés par Hydock (5) pour le magnésium

- 0,2% $K_2 Cr O_7$ + 0,2% $K NO_2$
- 1% Li F
- 1% d'huile de coupe hydraulub x

Le résultat de ces essais est donné fig. 8 et 9

L'inhibiteur au bichromate de potassium et l'huile de coupe donnent dans tous les cas une protection très efficace contre la corrosion sans altérer la surface des échantillons.

Les essais de corrosion sous tension ont finalement été faits dans une solution acqueuse contenant :

3% Na Cl
0,2% $K_2 Cr O_7$
0,2% $K NO_2$

le pH de cette solution était égal à 5.

2.2 Alliages utilisés et conditions d'essai :

Les alliages avaient les compositions suivantes :

	Mg	Al	Zn	Mn	Cu	Si
GA8-Z1	base	8,0	0,49	0,19	traces	traces
GA9	base	8,3	0,30	0,30	traces	traces

Le premier provenait d'une pièce matricée fissurée en service, les éprouvettes de corrosion sous tension y ont été prélevées suivant les sens long, travers long et travers court (le schéma des prélèvements est donné fig. 10).
Le second est un alliage de fonderie, les éprouvettes étaient soit à l'état homogénéisé (24 h à 400° refroidissement à l'air), soit à l'état homogénéisé vieilli (100 h à 160°).

Caractéristiques mécaniques

alliage	état	R_{hb}	$E_{0,2hb}$	A%	dureté Brinell $P/d^2 = 10$
GA8-Z1	long	32,7	20,1	15	67,5
	travers long	31,6	17,4	12	
	travers court	30,2	10,8	15	
GA9	homogénéisé	26,5	10,2	11	65
	vieilli	28	13,0	7	83

Les essais de corrosion sous tension ont été effectués en flexion à charge constante sur les éprouvettes préalablement décapées de la façon suivante :

- immersion 2 mn dans une solution à 18% $Cr O_3$
- rinçage à l'eau distillée
- séchage à l'air chaud

Quelques éprouvettes ont été mordancées avant essai afin d'étudier l'efficacité de la protection.

Les essais ont été faits en immersion continue; ils étaient interrompus par la rupture des éprouvettes ou après 30 jours d'essai,à l'issue desquels la charge de rupture résiduelle était comparée à celle d'éprouvettes corrodées en l'absence de contrainte.

2.3 Résultats

L'alliage de forge GA8-Z1 s'est révélé sensible à la corrosion sous tension dans tous les sens de prélèvement (fig. 11). Les limites de non rupture en 30 jours, dans les conditions d'essai, sont de l'ordre de 35% de la limite élastique pour le sens longitudinal et 50% de la limite élastique pour le sens travers court.

L'alliage de fonderie n'a pas donné de rupture pour des sollicitations inférieures à la limite élastique : toutefois, dans les deux états de traitement thermique on enregistre, sur les éprouvettes sollicitées à 8% de la limite élastique, une chute de 10 à 20% des caractéristiques mécaniques par rapport aux éprouvettes corrodées en l'absence de tension.

L'examen micrographique des éprouvettes a montré que les criques étaient principalement intergranulaires, sauf parfois au voisinage de la surface, sur l'alliage de forge.

Les quelques fissures décelées sur l'alliage de fonderie dans les deux états de traitement thermique étaient, la plupart du temps, transgranulaires.

Le mordançage des éprouvettes ne retarde pas l'apparition des ruptures (fig. 12); il semble même accélérer le phénomène à forte contrainte, sans doute parce que la couche mordancée se fissure sous charge. Il relève cependant un peu la limite de non rupture en 30 jours qui passe de 35 à 45% de la limite élastique.

3. CONCLUSION

Cette étude montre que les alliages de magnésium à 8% d'aluminium sont sensibles à la corrosion sous tension à l'état forgé ou matricé, mais non à l'état coulé; c'est un cas général pour tous les métaux.

Si l'on compare le comportement de ces alliages avec, par exemple, celui des alliages d'aluminium sensibles à la corrosion sous tension comme les aluminium-zinc et les aluminium-cuivre-magnésium, on remarque plusieurs différences :

- La sensibilité à la corrosion sous tension ne dépend pas du fibrage, on sait que cela n'est pas le cas des alliages d'aluminium dont la sensibilité ne se manifeste qu'en travers court. Cela évoque plutôt le comportement des alliages de titane. On peut remarquer, bien que le rapprochement soit hasardeux, que tous deux cristallisent dans le même système.

- Les fissurations peuvent être intergranulaires ou transgranulaires sur le même échantillon; dans le cas que nous avons étudié, ce comportement ne paraissait dépendre ni du traitement thermique, ni de la grosseur du grain.

- Enfin, la sensibilité à la corrosion sous tension peut être considérée comme relativement limitée, 35 à 50% de la limite élastique, alors qu'elle n'est que de 5 à 10% pour les alliages aluminium-zinc et aluminium-cuivre-magnésium non désensibilisés. C'est sans doute pourquoi les ruptures en service devraient être évitées plus facilement que sur ces alliages. Il suffirait en effet, et ce raisonnement découle à la fois des observations faites sur les pièces en service et sur les essais de laboratoire, de limiter avant tout les contraintes appliquées de façon permanente. Cela est sans doute valable pour toutes les pièces susceptibles de fissurer en corrosion sous tension, mais plus particulièrement pour les alliages de magnésium à cause de leur très faible limite élastique.

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(5) J.J. Hydock	Materials Protection May 1966

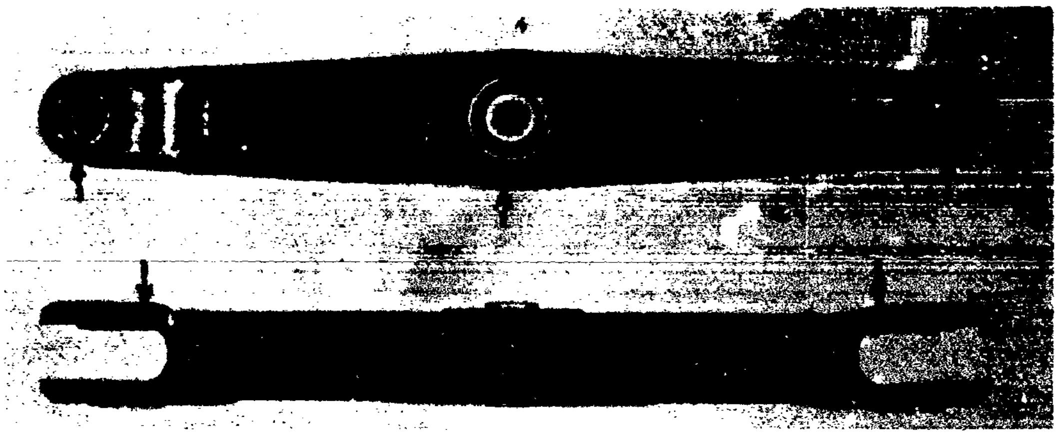

Fig. 1 Aspect d'une biellette de commande
→ localisation des ruptures

Fig. 2 Aspect d'une cassure ouverte → amorce de rupture
La rupture est due au pincement de la pièce au montage

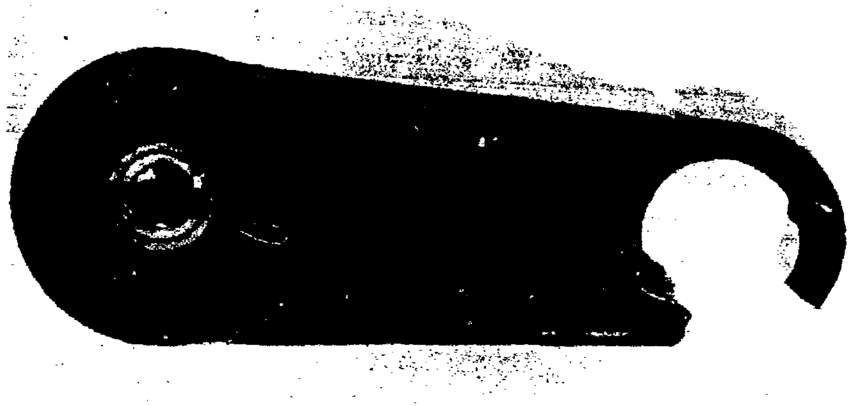

Fig. 3 Guignol cassé - la rupture est due au serrage excessif d'une bague de roulement

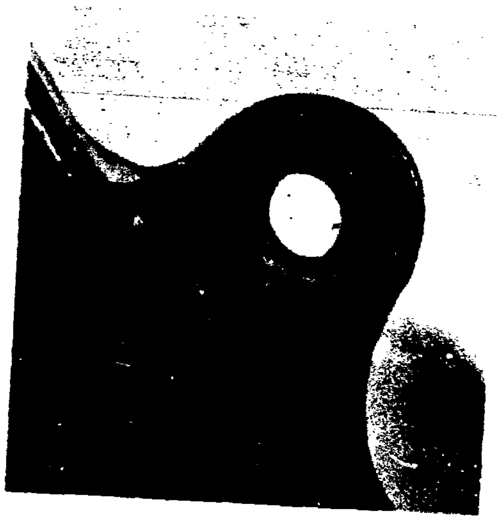

Fig. 4 Crique autour d'un trou bagué
Amorce sur une petite piqûre de corrosion

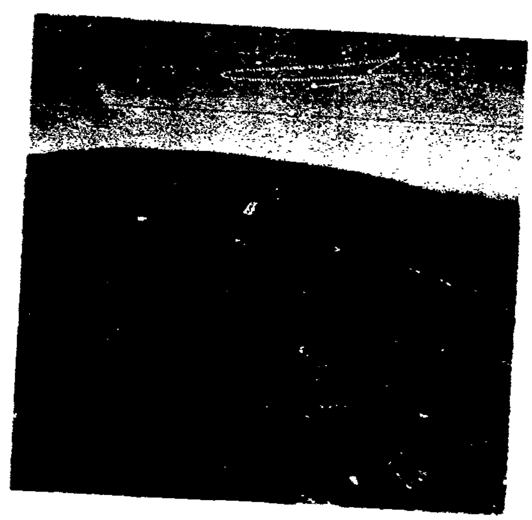

Fig. 5 Crique développée autour d'un poinçonnage

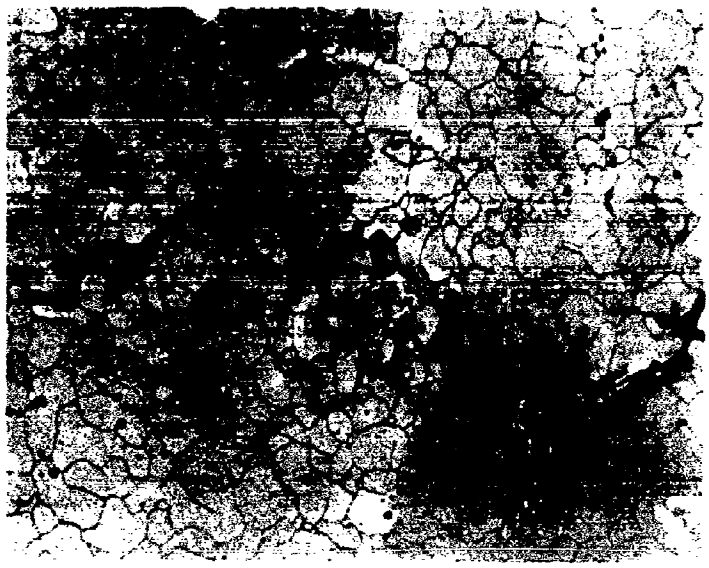

Fig. 6 Micrographie d'une fissure intergranulaire G = 300

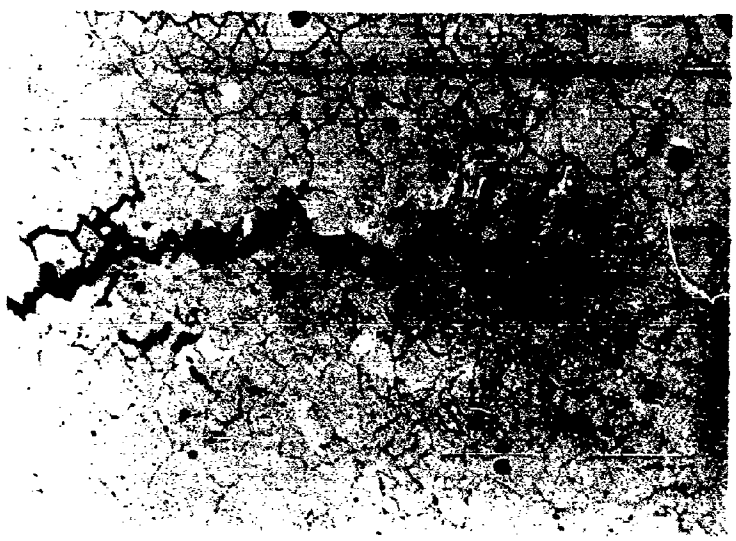

Fig. 7 Micrographie d'une fissure inter et transgranulaire G = 300

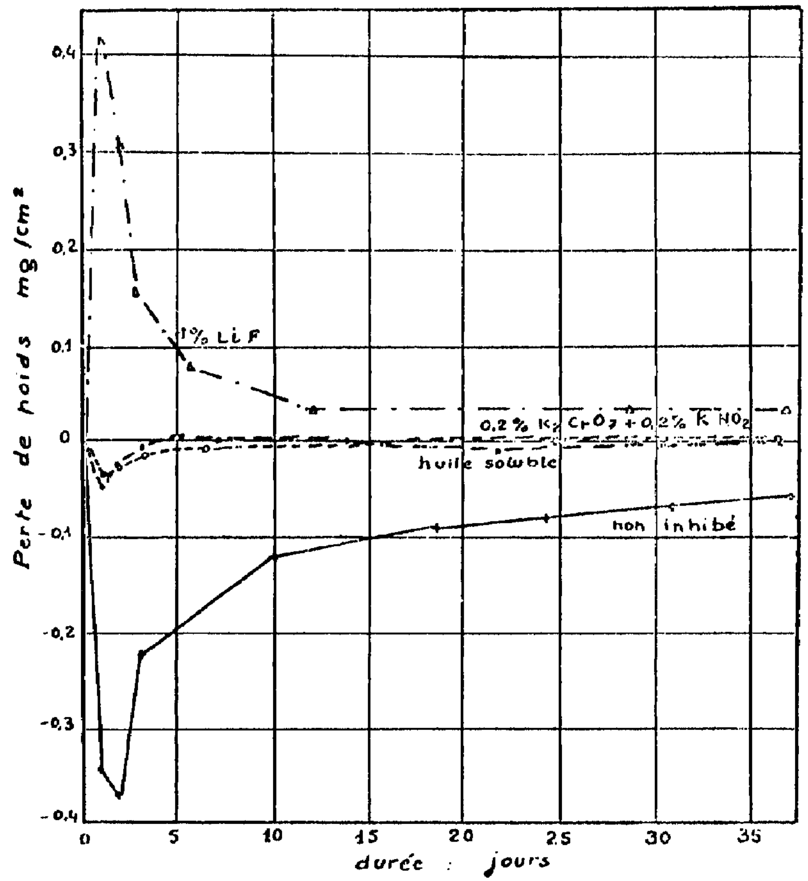

Fig. 8 Corrosion du GA8-Z1 dans l'eau à 3% de Na Cl
Effet des différents inhibiteurs

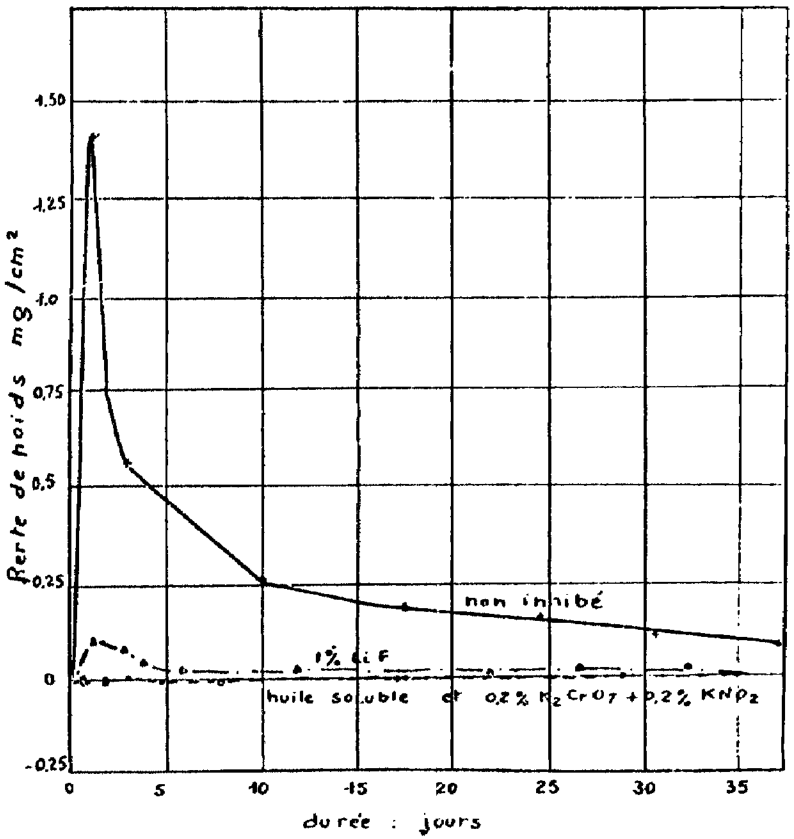

Fig. 9 Corrosion du GA8-Z1 dans l'eau du robinet
Effet des différents inhibiteurs

	R_{hb}	E_{hb}	A %
L	32,7	20,1	15
TL	31,6	17,4	12
TC	30,2	10,8	15

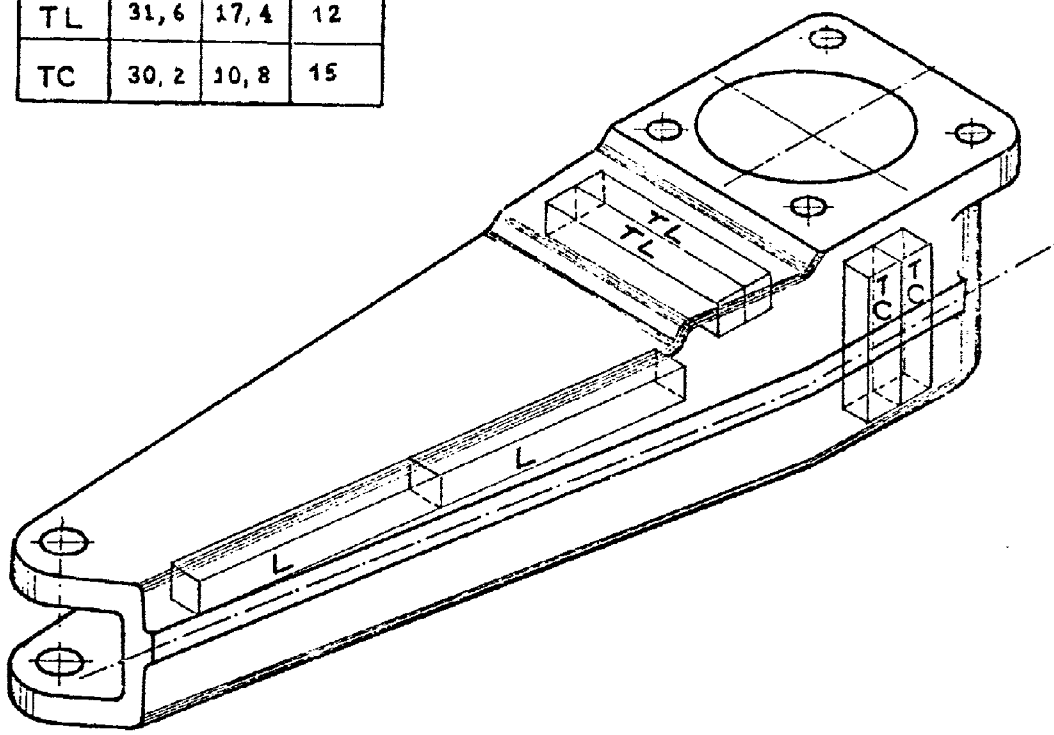

Fig. 10 Schéma des prélèvements dans la pièce matricée de GA8-Z1

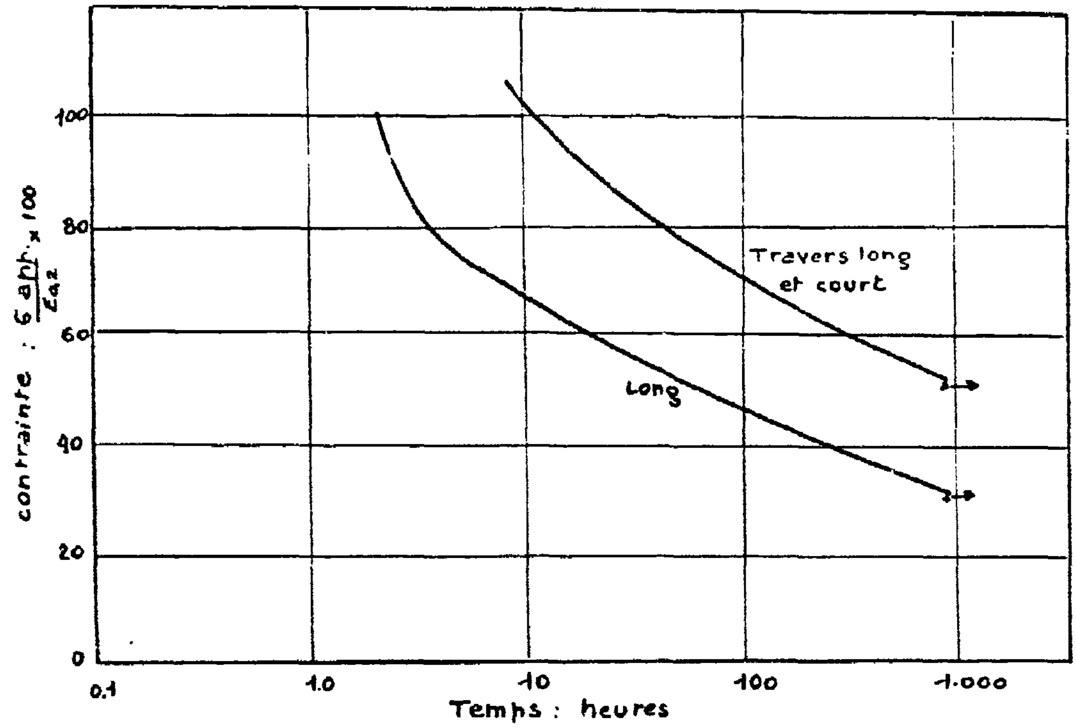

Fig. 11 Essai de corrosion sous tension - effet du sens de prélèvement

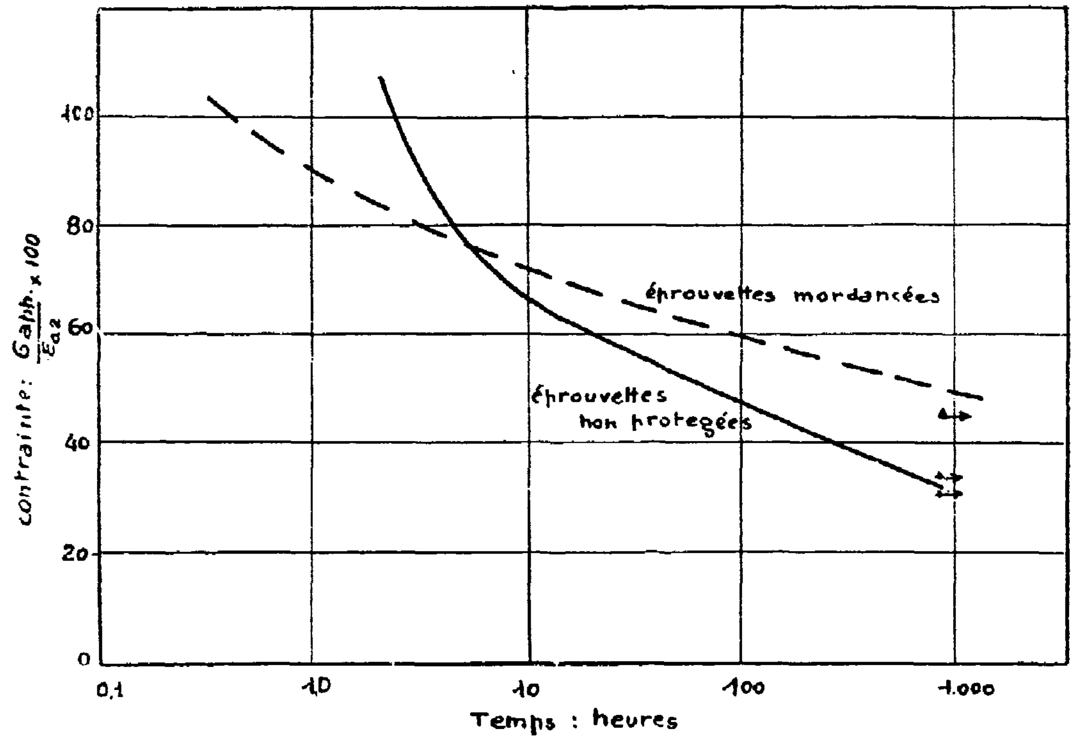

Fig. 12 Essais de corrosion sous tension - effet du mordançage.

THE RELATIONSHIP BETWEEN TEST RESULTS AND SERVICE EXPERIENCE

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SYMBOLS

K	Stress Intensity Factor
K_I	Stress Intensity Factor for Opening Mode
K_{Ii}	Initial Stress Intensity Factor for Opening Mode
K_{Iscc}	Stress Intensity Threshold Value for Stress-Corrosion Cracking
B	Thickness
TYS	Tensile Yield Strength
a_{cr}	Critical Crack Length

THE RELATIONSHIP BETWEEN TEST RESULTS AND SERVICE EXPERIENCE

SUMMARY

An attempt is made to correlate stress-corrosion data generated on smooth and precracked specimens in sodium chloride solution with the "safe-life" and "fail-safe" concepts used to provide reliability in high-performance structures. In the aircraft industry, most service experience with stress-corrosion cracking has occurred in high-strength aluminum alloy forgings and extrusions, and the specific "short-transverse" nature of the phenomenon has limited the usefulness of precracked specimen data to the designer. Should the next generation of aircraft materials exhibit even a moderate susceptibility to stress-corrosion cracking, which is independent of grain structure (e.g. titanium alloys), precracked specimen data will be vital to the designer in using the fail-safe concept. Since the materials engineer is responsible for establishing the stress-corrosion behavior of aircraft materials, testing methods are not always meaningful to the designer; thus, the possibility of developing a single specimen configuration to satisfy the needs of materials engineers and designers is explored.

1. INTRODUCTION

The safe use of materials in the design of high-performance vehicles requires cooperation between the materials technologist and the designer, emphasizing the evaluation and documentation by the materials man of properties that are meaningful to the designer. To achieve this, the materials man should not only check the correct materials and process instructions in a drawing but also acquaint himself with the design philosophy associated with the individual structural components. The fact that materials problems have been remedied in many cases by materials substitution or minor design changes demonstrates the need for good rapport between these engineering disciplines.

The design of airframe or missile structures incorporates a high degree of reliability and safety during the intended service life of the structure. The aircraft industry has generally recognized two philosophic concepts, "safe-life" and "fail-safe," as a means of providing structural reliability. The concept of safe-life design demands that no damage occur during the operational life of the structure. If damage does occur, a safety problem exists, and the service life of the structure is terminated. In contrast to safe-life, fail-safe design tolerates the initiation of unanticipated damage. This concept relies on the provision of a residual strength level in the damaged state that will not be exceeded before damage detection and repair are accomplished.

It is, therefore, the task of the materials technologist to recommend materials for aerospace components whose reliability and safety are based on safe-life or fail-safe criteria. Furthermore, this recommendation must be made with the lowest possible production cost, weight, and potential customer maintenance to provide the maximum product competitiveness and operational performance. These boundary conditions for choosing a material are particularly evident in the phenomenon of stress-corrosion cracking. Complete elimination of the inherent susceptibility of an alloy to stress-corrosion cracking without unduly sacrificing strength, toughness, fatigue life, fabricability, or other important properties always seems to require more fundamental knowledge about mechanisms than is known at the time.

From a metallurgical viewpoint, stress-corrosion tests are conducted for a number of reasons; a given reason frequently determining the type of test. A great deal of the confusion and derogatory comment concerning conventional stress-corrosion testing stems from not understanding why a particular test was selected. The materials technologist is constantly seeking an accelerated test to discriminate between the stress-corrosion susceptibilities of high-strength alloys while the designer tends to use the same test data for assistance in selecting a material for service in a given environment, the designer also expects to make a reasonable forecast of its behavior under service conditions. Attempts to seek a compromise test method have further complicated some techniques but also have provided impetus for the development of precracked specimens as a quantitative technique for assessing the stress-corrosion susceptibility of high-strength alloys.

2. STRESS-CORROSION RESULTS FROM SMOOTH SPECIMENS

Before attempting to put into perspective for the designer the meaning of stress-corrosion data generated from smooth specimens, the salient points about conventional test techniques will be briefly described. The choice of specimen configuration is usually determined by the form in which the metal is available. Table I relates the various specimen types to material form. In the selection of the method of loading the specimen, the metallurgist can attempt to simulate the type of stress the part might encounter in service. Since many stress-corrosion failures are probably the result of residual and installation stresses, tests employing constant deflection are probably the most realistic. Constant load tests may simulate more closely failure from applied or working stresses. Table II lists some types of stress that prevail in service and some recommendation as to whether constant deflection or constant load would more nearly duplicate the stress conditions. Specimens should be stressed in more than one direction with respect to the rolling direction. Behavior when stressed in the short-transverse grain direction is especially important in the case of high-strength aluminum alloys.

TABLE I. Relationship of Specimen Types and Material Forming[1]

Form in Which Material is Available	Possible Types of Specimens
Sheet	Bent beam, preform, tension specimen, U-bend
Plate (less than 2 in. thick)	C-ring, bent beam, tuning fork
Plate (more than 2 in. thick)	Short-transverse tension specimen
Bar (depending on thickness or diameter)	C-ring, tuning fork transverse tension specimen
Tubing	C-ring (Battelle system of internal pressure)
Wire	Tension specimen, loop

Specimens for Simulating Specific Conditions

1. Preformed specimens for simulating residual stresses
2. Welded assemblies
3. Interference rings for simulating pressed in bushings, fasteners, etc.

TABLE II. Relationship Between Service Stresses and Stressing Methods[1]

Source of Sustained Tension in Service	Stressing Method Most Applicable
Residual stresses	
Quenching, Forming	Constant deflection
Misalignment	Constant deflection
Interference bushings	
Rigid	Constant deflection
Flexible	Constant load
Flareless fittings, Clamps	Either, but constant deflection probably better
Hydraulic pressure	Constant load
Deadweight	Constant load
Faying surface corrosion	Constant load

Environment is a very important aspect of testing methods for stress-corrosion cracking. Laboratory stress-corrosion tests are usually closely controlled and conducted in such manner as to accelerate the stress-corrosion process and to duplicate the type of failure experienced in a service environment. For the most part, stress-corrosion testing in the atmosphere is used to correlate laboratory tests and to determine how the alloy will behave in a realistic environment. Accelerated tests are naturally more useful if the data correlate with service experience and environment. In the absence of a correlation, the tests may still be used for screening purposes. Most workers in the stress-corrosion field use a test period that is based on the particular test method, environment, material, convenience or space, and the end use of the data. Test periods may range from one day to three years or until the material fails, but usually the duration is long enough to give reasonable assurance that the material will perform satisfactorily in a given environment.

Stress-corrosion cracking characteristics of a material are usually obtained by subjecting smooth specimens to several applied tensile stresses. The data are generally presented graphically with stress level plotted against time-to-failure of the specimen (Figure 1). The resulting curve exhibits a threshold stress below which test specimens do not fail under the prevailing testing conditons. Table III summarizes the highest sustained tensile stress (ksi) at which several aluminum alloy specimens at different orientations did not fail in the 3.5% NaCl alternate immersion test. Stress-corrosion data generated in this manner provide comparisons of the resistance of high-strength materials on the basis of laboratory tests. If there has been a useful correlation of the laboratory test with atmosphere exposure tests, the threshold stress should indicate to the designer that the part should not be continuously stressed in tension above this value in the specific orientation considered.

TABLE III. Guidelines for Comparing Resistance to Stress Corrosion of Various Alloys and Products[2]

Summary of Highest Sustained Tensile Stress (ksi) at Which Test Specimens of Different Orientations Did Not Fail in the 3½% NaCl Alternate Immersion Test

Alloy and Type of Temper	Test Direction	Plate	Rolled Rod and Bar	Extruded Shapes Section Thickness 0.25 - 1 in.	Extruded Shapes Section Thickness 1.25 - 2 in.	Hand Forgings
2011-T3	L	–	25	–	–	–
2011-T3	T	–	10	–	–	–
2011-T8	L	–	>40	–	–	–
2011-T8	T	–	>30	–	–	–
2014-T6	L	45	45	50	45	30
2014-T6	LT	30	–	27	22	25
2014-T6	ST	7	15*	–	7	7
2024-T3, -T4	L	35	30	>50	>50	–
2024-T3, -T4	LT	20	–	37	18	–
2024-T3, -T4	ST	7	10*	–	7	–
2024-T6, -T8	L	>50	>47	>60	>60	–
2024-T6, -T8	LT	>50	–	50	50	–
2024-T6, -T8	ST	43	>43*	–	16	–
2219-T8	L	>40	–	>35	>35	>38
2219-T8	LT	>38	–	>35	>35	>38
2219-T8	ST	>38	–	–	>35	>38
7075-T6	L	50	50	60	60	35
7075-T6	LT	45	–	50	32	25
7075-T6	ST	7	15*	–	7	7
7075-T73	L	–	>50	–	–	>50
7075-T73	T,ST	–	>47*	–	–	>48
7079-T6	L	>55	–	>60	>60	>50
7079-T6	LT	42	–	50	35	30
7079-T6	ST	7	–	–	7	7
7178-T6	L	55	–	65	65	–
7178-T6	LT	38	–	45	25	–
7178-T6	ST	7	–	–	7	–

* Ratings are for transverse specimens machined from round or square bar stock.

3. STRESS-CORROSION RESULTS FROM PRECRACKED SPECIMENS

A major breakthrough in testing methods for stress-corrosion cracking occurred as the result of a very expensive experience with titanium alloys. The use of certain titanium alloys was contemplated for the structural components of future high-speed commercial aircraft, and extensive conventional testing using smooth specimens indicated their immunity (or extremely high resistance) to sea water at room temperature. However, it was discovered[3] that an identical environment would dramatically propagate an existing natural crack in the same alloy at extremely low applied stresses.

The prerequisite of the prior existance of a sharp crack to measure the stress-corrosion characteristics of titanium alloys has led to the presentation and analysis of the test data by linear elastic fracture mechanics methods. These methods have provided a quantitative treatment of both subcritical cracking and brittle fracture behavior in real structures during the last decade. The single parameter K characterizes the elastic stress field in the leading edge of a crack. Relationships between K, stress, and crack size for a number of different geometries are available and provide the information for predicting fracture behavior from one geometry or stress to another.

Specimen configurations, suitable for studying the fracture mechanics of stress-corrosion cracking, have been discussed by Brown[4] in a detailed review of the applications of fracture mechanics to stress-corrosion cracking. All specimens used in stress-corrosion testing were developed originally to evaluate the unstable fracture toughness of metals and alloys. Consequently, the validity requirements follow closely the recommendations of Brown and Srawley[5] for plane strain fracture toughness testing. Testing methods may be conveniently divided into three categories by reference to the relationship between plane strain stress intensity factor for opening mode K_I and the crack length. This intensity of the crack tip field can be made to increase, remain constant, or decrease as the crack propagates. Titanium alloys are generally evaluated using the former relation between K_I and crack length, aluminum alloys using the latter relation, and steels either using the former or latter relationship. The constant K_I-type test has seen limited use during academic studies.

The single edge notched specimen (Figure 2) can be used to illustrate a typical testing procedure for titanium alloys. The plane strain fracture toughness K_{Ic} is usually determined for reference by loading the specimen to failure in air by three- or four-point bending. To evaluate stress-corrosion characteristics, the fatigue cracked notched bend specimens are sustain loaded in cantilever bending in the environment to the selected initial stress intensity factor K_{Ii} levels below K_{Ic}. The time required for failure is plotted against the corresponding initial stress intensity level (Figure 3). For titanium alloys, a threshold-stress-intensity-factor level exists and has been termed K_{Iscc}. This threshold is the minimum value of K_I at which stress-corrosion crack growth occurs after a certain time in a specific environment. Specimens subjected to K_{Ii} levels above K_{Iscc} usually fail in sea water or sodium chloride solution within 20 min. In a similar environment, most steel and aluminum alloy specimens would exhibit considerably slower crack growth rates, especially when K_{Ii} is close to the threshold K_{Iscc}. As a result, steel and aluminum specimens must be closely inspected for evidence of crack growth at the end of the test period.

Hyatt[6] has recently studied the use of precracked specimens to determine the stress-corrosion susceptibility of high-strength aluminum alloys. He has developed a test method that indicates the relationship between stress intensity factor K_I and stress-corrosion crack growth rate. A double cantilever beam specimen (Figure 4) was used for the stress-corrosion test. Some attractive features of this specimen are a capability to be self-stressed by a bolt or wedge in the short transverse grain direction and elimination of fatigue precracking to provide the artificial stress raiser.

To perform the test, the specimen is fixed in a vise or other holding device, and the environment is applied to the tip of the machined notch. The arms of the specimen are deflected (by turning the bolt or driving the wedge) until a natural crack "pops-in" from the notch. The crack opening displacement v is measured along the line of load application and maintained at this level for the duration of the test. The specimen is immediately subjected to predetermined environmental conditions, and crack length is monitored as a function of time elapsed from pop-in. The overall result of this procedure is to cause the stress intensity factor to decrease as the crack extends under the influence of an aggressive environment. The slope of the crack length versus time curve at any crack length will provide the crack growth rate, while the use of an equation converts the crack length to stress intensity factor K_I. This provides the curve of stress intensity factor as a function of crack growth rate due to the environment. The K_I at crack arrest (or at some extremely low crack-extension rate) will indicate the threshold stress intensity level for stress-corrosion cracking. Figure 5 illustrates data taken in this manner on three commercial aluminum alloys, which were intermittently wetted with 3.5% sodium chloride solution.

Novak and Rolfe[7] were the first to use a modified version of the wedge-opening-load (WOL) specimen, originally developed by Manjoine[8], for stress-corrosion testing of steels. By stressing the WOL specimen with a bolt, these workers also studied a crack arrest-type specimen (i.e. the stress-intensity factor decreased with increasing crack length). However, they did encounter some difficulty in preventing the arms of the specimen from breaking off and in keeping the stress-corrosion crack in plane. It was also necessary to fatigue precrack the steel specimens because they were too tough to pop-in manually. As mentioned previously, high-strength aluminum alloys can readily be precracked by pop-in methods, and the popular short transverse grain direction path of stress-corrosion cracks ensures a planar crack for accurate fracture mechanics analysis.

To summarize, the following types of stress-corrosion data are available to the designer:

a. A threshold tension stress (in ksi) below which test specimens did not fail (or stress-corrosion cracks did not initiate) under the prevailing test conditions—from smooth specimen geometries.

b. A relationship between stress-intensity factor (which is a function of stress and flaw size for a given geometry) and stress-corrosion-crack growth, and also a threshold-stress-intensity-factor level below which stress-corrosion cracks will not propagate in the prevailing environmental conditions of the test—from precracked specimens.

Having endeavored to put into perspective the stress-corrosion cracking results available to the designer, the next stage is to correlate these data into design terms for each major alloy system.

4. CORRELATION OF STRESS-CORROSION DATA INTO DESIGN TERMS

Some knowledge of design terms is an obvious prerequisite to understanding the type of stress-corrosion data that the designer needs to ensure structural reliability. The safe-life and fail-safe terms discussed previously were rigid definitions, which found an original application as a safeguard against the fatigue phenomenon. The fail-safe concept, in particular, has received expanded definition to encompass backup systems for safe-life components, electronic devices, et cetera.

Smooth specimen threshold data may only be meaningful in the design of safe-life structural components because they only stipulate that under special testing conditions no failures occurred at stresses below the threshold stress. Smooth specimen data do not distinguish between initiation and propagation of cracks and thus do not aid in establishing crack inspection intervals. An important aspect of fail-safe design is the ability to detect cracks early. Crack propagation rate information about the actual material in the anticipated service environment (including load and temperature) is the primary form of data required. The residual strength of the structure is also important because fail-safe structure is designed to have reasonable residual strength in the damaged or cracked condition.

In the fail-safe fatigue concept, crack growth is predicted using linear cumulative crack-growth hypotheses and data derived by stress intensity factor concepts. The latter have been used to analyse stress-corrosion test results conducted on precracked specimens, and the data offer a plausible opportunity for use as a means of determining nondestructive inspection criteria and the residual strength of cracked components subjected to aggressive environments.

Designers strive to attain fail-safe reliability in the construction of high-performance aircraft. When fatigue performance of a material does not permit economic surveillance of damage rate or if the component is not easily accessible for inspection, it may be designated safe-life and be scheduled for replacement well in advance of anticipated fatigue damage. However, such components are subjected to full-scale tests under realistic (but accelerated) loading spectra to provide information for replacement intervals.

During stress-corrosion studies, testing of complete components is rather impractical, and the time for simulation of anticipated sustained surface tension stresses and service corrosive environment cannot be realistically shortened as is claimed for the case of a fatigue evaluation. Laboratory stress-corrosion data that show favorable comparison with service experience are available, but in many cases the smooth specimen data cannot be collated when tests are conducted at different laboratories. Because of these situations, it is perhaps understandable that the designer is reluctant to freely discuss the direct application of the safe-life reliability concept as a safeguard against stress-corrosion cracking.

In the aluminum alloy system the occurrence of stress-corrosion cracking has been attributed to sustained residual and assembly tension stresses acting specifically in the short-transverse direction with respect to the grain-flow pattern of forgings and extrusions. Having identified the cause of cracking, remedial action has been taken by the materials engineer to reduce residual stresses due to heat-treatment, forming, straightening, and machining operations and by the fabricator to control installation-induced stresses. The designer, recognizing the overwhelming preference of the phenomenon to act in the short transverse (ST) grain direction (Table III), might feel absolved from responsibility because sustained tensile loads are not normally designed to act in this grain direction in extrusions and forgings. However, the exact location of the short transverse grain direction with respect to design loads might be difficult in complicated components.

When all feasible methods of reducing sustained tensile stresses acting in the short transverse direction have been applied to the part, a comparison is made between the remaining stresses and the stress-corrosion threshold determined during smooth specimen tests. To protect against crack initiation, the designer must ensure that the sum of the sustained residual, assembly, and design tensile stresses (including design safety factors) does not exceed threshold stress for the material or components. There is an excellent chance that the stress-corrosion threshold will be low in the short transverse direction (e.g. 7 ksi for 7079-T6, 7075-T6, and 2024-T3, -T4 in Table III) because of optimization of other essential properties. If the designer considers that the reliability of the part is marginal, he might either select one of the more resistant alloys and tempers such as 7075-T73 (and perhaps take a severe weight penalty) or improve the structural reliability of the more susceptible but stronger alloys.

The only "second line of defense" reliability for susceptible alloys is to introduce surface compressive stresses to the component (by peening) and to exclude the corrosive environment. In this latter respect, there is always the hope that a suitable protective coating can be applied that will protect the stressed part from stress-corrosion cracking. It has been demonstrated that anodizing, cladding, painting, et cetera does provide considerable protection against the natural initiation of cracks in aggressive service environments; however, both the materials engineer and the designer would prefer a better component reliability criterion than the application of a surface coating. If flaws and cracks can be so effective in causing loss of structural integrity, questions immediately arise as to their early detection and the manner of their origin and development. Scratches, pits, flaws, and other defects are characteristic of the delivered raw stock, fabrication history of the part, and the customer's use of the product.

Prior to introduction of precracked specimens in stress-corrosion testing, the data could not be analysed and used by the designer to predict the behavior of cracked components. This situation and the realization that some painted parts did fail by stress-corrosion cracking has prompted the replacement of susceptible alloys by highly resistant alloys for application in critical components. In the case of 7075-T73, the high resistance to stress-corrosion cracking is accompanied by a 15% reduction in tensile strength compared with 7075-T6.

The increasing demands on the designer to produce more efficient structures have resulted in the need for material of higher conventional strength properties. He is naturally reluctant to take a 15% loss in strength to obtain total immunity in a grain direction that does not purposely contain design loads anyway. The efforts of the materials engineer to provide stronger aluminum alloys with sufficient stress-corrosion resistance have been underscored by the inability of traditional stress-corrosion test techniques to screen the candidate alloys.

The tensile properties of 7000 series aluminum alloy (the measurement of which few engineers doubt) have been competitively increased to meet new technical design requirements with the corresponding claim by the alloy producers that the materials also have a high resistance to stress-corrosion cracking. To substantiate these claims, the alloys are subjected to standard stress-corrosion tests (the validity of which everyone questions), and generally all competitive products survive the testing environment. The most frustrating aspect of this situation is that rather than entertain the idea of a new approach to testing methods, alloy producers and materials technologists have tended to make the existing testing conditions more severe by extending the test period,

plastically deforming specimens, and in some cases adding concentrated acid to the test solution! The results of these tests will, of course, be much less meaningful to the designer than data obtained in environments where a satisfactory correlation with service behavior had been established.

The materials engineer has benefited most by the development of test methods for aluminum alloys that use precracked specimens; the designer has only benefited indirectly. Some advantages of the new technique and quantitative data to the materials engineer are

a. Design of the double cantilever beam specimen is extremely simple. It is inexpensive to prepare and does not require fatigue precracking to produce an artificial stress raiser. It can be self-stressed in the susceptible short transverse grain direction and is portable. Crack-length measurements are made using a rule and a magnifying glass. The low cost of preparing the specimen is the key to its eventual widespread use.

b. Although the specimen is very new compared with conventional smooth specimens, early indications are that the virtual elimination of the crack-initiation stage and certain geometry effects (which influence total time-to-failure on smooth specimens) enable satisfactory comparison of data generated at different laboratories.

c. The rating of stress-corrosion susceptibilities using double cantilever beam specimens agree with the established trends based on smooth specimen threshold data. Where some discrepancies occur, the precracked data has correctly forecast the service experience with the alloy. A good example is the case of commercial alloys 7075-T6 and 7079-T6. Smooth specimen data indicate a threshold of 7 ksi for both alloys in 3.5% sodium chloride-alternate immersion and in sea coast atmosphere (Figure 1). Users of these two alloys agree unanimously that 7079-T6 has the worst service record. The precracked data in Figure 5 indicate a similar K_{Iscc} value for each alloy but a different crack-growth rate that increases as the stress intensity level increases. For example, if cracks had initiated in both alloys and a stress intensity factor was calculated as 10 ksi $\sqrt{in.}$ (from crack geometry and residual tensile stress considerations), the crack growth rate would be 10 times faster for the 7079-T6 alloy.

d. Tests conducted with the double cantilever beam specimen have provided very rapid and discriminatory stress-corrosion data about developmental high-strength aluminum alloys and aided in selection of heat-treatment conditions, which impart the best stress-corrosion properties to these alloys[9]. Between the very slow crack growth rate of the highly resistant 7075-T73 alloy and the extremely rapid rate of growth associated with 7079-T651, there are six logarithmic cycles to characterize the competitive products of aluminum alloy manufacturers (Figure 5). Furthermore the alloys can be effectively rated in terms of crack growth rate after only about one week. Thresholds take substantially longer to establish especially in the relatively immune alloys.

e. A similar double cantilever beam specimen is being used by the fundamental metallurgist to investigate the mechanism of stress-corrosion crack growth. Speidel[10] has used the specimen to study the influence of halide ion concentration, electrochemical potential, and viscosity on the crack velocity in commercial aluminum alloys and Brown et al.[11] to investigate the solution chemistry within stress-corrosion cracks. Thus, the material engineer has a direct line of communication with important mechanism and alloy development studies in aluminum alloys where the data are directly comparable to data generated on commercial alloys.

f. The double cantilever beam specimen could form part of a quality control procedure for aluminum alloy plate, forgings, and extrusions. Material could be accepted on the basis of stress intensity factor versus crack growth rate data and K_{Iscc} threshold data. The precracked specimen technique would identify susceptible alloys in about one week.

g. Results of laboratory tests to measure susceptibility to stress-corrosion cracking should closely simulate the behavior of the component in service. A reasonable correlation has been developed for smooth specimens, but generally specimens must be exposed outdoors for up to three years. These tests suffer the fate of smooth specimen laboratory tests in that poor control of initiation phenomena causes scatter in the results. The double cantilever beam specimen seems ideally suited for service simulation tests since crack growth rate and threshold K_{Iscc} data are not influenced by initiation phenomena, and it should be easy to match the crack-growth-rate characteristics of the service environment in the laboratory. Figure 6 shows data developed by Hyatt[6] on 7075-T651 by subjecting the double cantilever beam specimen to different environmental conditions.

Because of the recent development of precracked specimen testing methods and, more important, the recent public scrutiny of data, this technique is not meant to completely supplant smooth specimen testing in aluminum alloys. Rather, stress-corrosion crack growth rate and K_{Iscc} data should be considered a valuable addition to smooth specimen threshold data in the same way that fatigue crack-growth-rate data supplements the standard S-N fatigue information. The direct benefit to the designer is the generation of quantitative data that is capable of predicting the behavior of components in service. The unique feature of the precracked double cantilever beam specimen is that it is suitable for the measurements of stress-corrosion characteristics from the fundamental to the engineering design level and is, therefore, a prime candidate for a standard in aluminum studies. Evaluation of the stress-corrosion sensitivity of welded specimens has not been pursued vigorously mainly because of the difficulty in locating the precrack in specific areas of the weld.

Titanium alloys had been considered immune to stress-corrosion cracking because of the high resistance of smooth specimens stressed to high percentages of their tensile yield strength in sea water at room temperature. However, in the presence of an artificial stress raiser, such as a fatigue crack, the crack-propagation resistance in the same environment was dramatically reduced for certain titanium alloys. These alloys were equally sensitive in all grain directions and thus the designer could not design his sustained service loads around the phenomena as was the case in aluminum alloys. Furthermore, the reduction in the load-carrying capability of cracked specimens in aqueous chloride was so great (70% for titanium 8Al-1Mo-1V alloy) that the phenomenon seriously undermined the other attractive properties of titanium alloys that had prompted their original selection for high-speed, high-performance aircraft.

The primary objective of the materials engineer was to learn how to increase the crack propagation resistance of high-strength titanium alloys, and it became imperative that the designer receive test data that would ensure structural reliability by fail-safe design. Data were required about the residual strength of damaged titanium structure so that the designer might safely use alloys that are moderately susceptible.

Figure 7 contains stress-corrosion data generated at Naval Research Laboratories[12] and The Boeing Company[13] in 3.5% sodium chloride solution. Test methods have been satisfactorily compared at both laboratories. These techniques, based on linear elastic fracture mechanics principles, give the best definition of stress-corrosion cracking behavior of these alloys in terms of K_{Iscc}. The spectrum of data indicates that suceptibility is strongly influenced by metallurgical variables such as composition, processing, heat-treatment condition, and microstructure. No attempt has been made to identify the individual compositions, heat-treatments, et cetera associated with each K_{Iscc} threshold; these data are readily available in the referenced literature.

An optimum relationship between K_{Iscc} and tensile yield strength is indicated for commercial and near-commercial titanium alloys. The corresponding relationship between fracture toughness and tensile yield strength is also shown. Fracture mechanics tests require that maximum mechanical constraint exist at the crack tip to ensure that conditions for valid plane strain measurement are attained. It has been specified[14] that conditions for plane strain and minimum K_{Iscc} threshold values are definitely met if the thickness of the specimen conforms to Eq. (1) as follows[15]:

$$B \geq 2.5\,(K_{Iscc}/TYS)^2 \tag{1}$$

where: B = specimen thickness

K_{Iscc} = plane strain stress intensity threshold value for stress-corrosion cracking

TYS = tensile yield strength

The thickness requirement for maximum mechanical constraint effectively limits the reliable determination of minimum K_{Iscc} threshold values to materials of K_{Iscc}/TYS ratios below 0.55[12] and 0.45[13]. These maxima constraint ratios are plotted on Figure 5. To determine whether K_{Iscc} values above these ratios are accurate, further tests would be required with specimens of increased thickness.

The K_{Iscc} values obtained for a particular specimen geometry can accurately predict conditions for stress-corrosion cracking in other specimen configurations provided the condition for plane strain is met. Goode et al.[16] have compared the stress-corrosion response of various alloys by calculating the depth of a long surface crack (at least 10 times as long as it is deep) that would be sufficient for stress-corrosion-crack propagation at yield point stress levels. For these conditions, the appropriate equation reduces to Eq. (2) as follows:

$$a_{cr} = 0.2\,(K_{Iscc}/TYS)^2 \tag{2}$$

where a_{cr} is the shallowest crack expected to propagate a stress-corrosion crack. Four arbitrary values for a_{cr} are included in Figure 7. Such a plot can be used as follows[4]. If an investigator knows that his inspection system for a given structure can detect all long surface cracks deeper than 0.01 in., then the titanium alloy he would select would be an alloy have a K_{Iscc} above the 0.01-in. line. Conversely, if the K_{Iscc} and tensile yield strength of a material are known, the equation may be used to estimate the maximum tolerable flaw size. Substitution of anticipated design stress in terms of "percentage TYS" in the latter equation will generate a new series of a_{cr} lines of lower slope in Figure 7.

For a specific flaw subjected to a stress intensity less than the K_{Ic} value but above the K_{Iscc} value, crack growth due to stress corrosion will occur in 3.5% sodium chloride until the flaw size-stress condition corresponding to K_{Ic} stress intensity is reached. The value of this type of information to the designer is immediately evident although, to design against stress-corrosion cracking, a detailed knowledge of stresses in the critical regions of complicated structures is necessary. If critical flaw sizes are too small and below reliable detectability limits (especially if stresses are not known accurately), the designer will be forced to discuss with the materials engineer those trade-offs he is willing to give for certain types of structural application.

Smooth specimen tests are conducted to evaluate the hot-salt stress-corrosion susceptibility of titanium alloys. The specimen, method of stressing, and analysis of data are similar to the description in Section 2 except that the tests are carried out at elevated temperatures.

Low alloy steels, quenched and tempered to tensile strength levels up to 300 ksi, are employed in a number of critical structural components. These steels are usually selected because of other properties besides stress-corrosion resistance, but precautions are taken to alleviate poor stress-corrosion characteristics. Components are often plated and painted to exclude the environment.

The use of precracked specimens to supplement smooth specimen data has thrown some light on the crack propagation characteristics of high-strength steels. As mentioned in Section 3, steels may be evaluated by a self-stressed WOL-type specimen[7] or by a single-edge notched specimen. Propagation rates are slow compared with those of titanium alloys, and thus, in the latter specimen type, crack growth rate may be monitored. Carter[17] has reported that additions of silicon up to 2.15% to 4340 steel did not increase the threshold stress-intensity parameter K_{Iscc} in the 280- to 300-ksi tensile strength range (Figure 8). However, the stress-corrosion crack velocity was significantly retarded when the silicon content exceeded 1.5% (Figure 9).

Crack propagation data aid in interpreting some of the unnotched bent beam tests, which have been conducted on many steels. Also, if all properties of a high-strength steel are similar for a particular application, the designer may have the added opportunity to select the alloy with the slowest propagation rate for extra reliability in the safe-life design of the part.

5. CONCLUSIONS

To assess the usefulness of stress-corrosion data to the designer, the materials engineer (such as the author) has considered that structural reliability is generally obtained by either a fail-safe or safe-life design concept. Data from tests conducted on precracked specimens offer an opportunity for the designer to establish inspection criteria and to calculate the residual strength of cracked components (i.e. useful in fail-safe design). Data determined on smooth specimens indicate that the material will not fail by stress-corrosion cracking (or maybe initiate a stress-corrosion crack) below a threshold stress level (i.e. useful only in safe-life philosophy). The designer is usually reluctant to design to safe-life unless tests have been conducted on full-scale components. Such tests are apparently more difficult to perform in stress corrosion than fatigue.

It is suggested that the noticeable occurrence of stress corrosion in aluminum alloys in the short transverse grain direction has resulted in the designer tolerating surface treatments of components to offset poor stress-corrosion resistance and provide structural reliability. As a result, the quantitative data generated from precracked specimens of aluminum alloys has only been directly beneficial to the materials man. It is emphasized that as soon as the stress-corrosion cracking phenomenon becomes no respector of grain direction, quantitative data of the type resulting from precracked specimens will be an important asset to designer information.

Precracked specimen testing for stress-corrosion cracking is relatively new compared with the conventional smooth specimen techniques. For this reason, it is intended that data from the different techniques complement each other. The use of both techniques offers an independent evaluation of initiation and propagation phenomena. The contribution of initiation to the total stress-corrosion process has not been satisfactorily resolved and as far as the precracked specimen philosophy is concerned, if artificial flaws will not grow, then whether they initiate naturally or unnaturally is of minor concern.

The designer is also concerned about the actual testing conditions used by the materials engineer and specifically whether the threshold stress et cetera will be valid for service environment. It is proposed that specimen standardization will eliminate much of the argument between the materials engineer and the designer and provide an excellent line of communication between these disciplines. The materials engineer is now providing qualitative data on smooth specimens and quantitative results on precracked specimens for use by the designer. The designer, realizing that he might be forced to use materials that are moderately susceptible, should reciprocate by indicating whether the data are sufficient, what combinations of properties are required, and what trade-offs he is willing to give for certain types of structural application.

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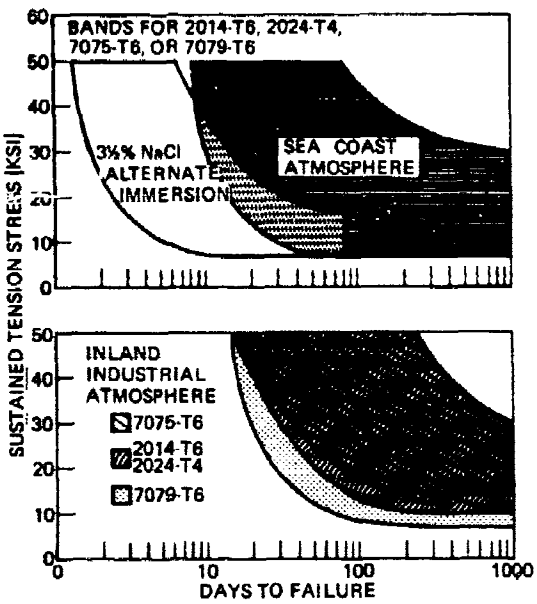
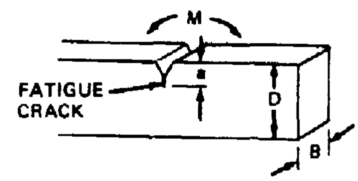


Figure 1. Comparison of the Performance of Short Transverse Specimens in Several Environments (2)



$$K_I = \frac{4.12\, M \sqrt{\frac{1}{\alpha^3} - \alpha^3}}{BD^{3/2}}$$

$$\alpha = 1 - \frac{a}{D}$$

Figure 2. Cantilever-Beam Specimen and Associated Formula

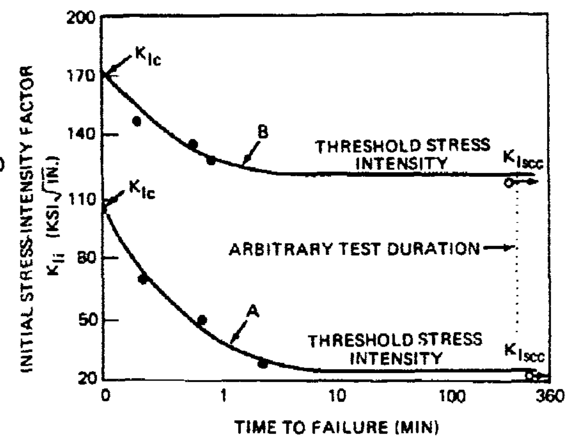
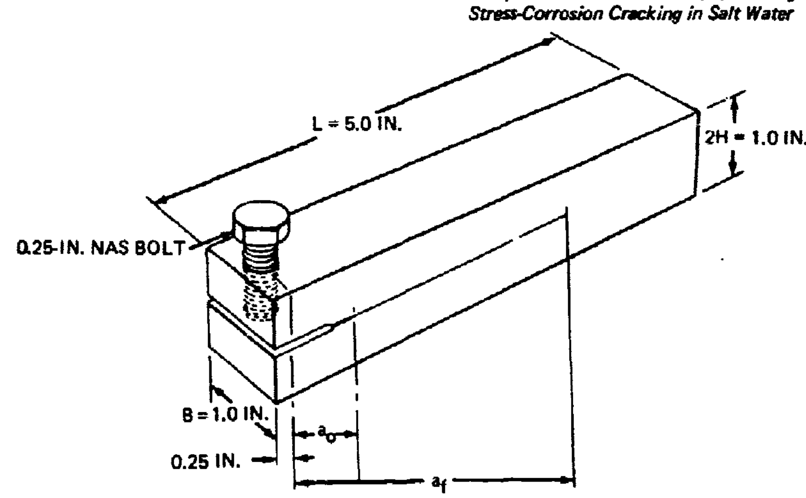


Figure 3. Typical Stress Corrosion Characterization for Titanium Alloy With Low Resistance (A) and High Resistance (B) to Stress-Corrosion Cracking in Salt Water



a_o = CRACK LENGTH AT "POP-IN"

a_f = CRACK LENGTH AT ARREST

$$K_I = \frac{EvH}{4} \frac{\left(H\left[3(a+0.6H)^2 + H^2\right]\right)^{1/2}}{\left[(a+0.6H)^3 + H^2 a\right]}$$

WHERE:

E = MODULUS OF ELASTICITY

v = CRACK OPENING DISPLACEMENT AT BOLT LINE

Figure 4. Double Cantilever Beam Specimen and Associated Formula

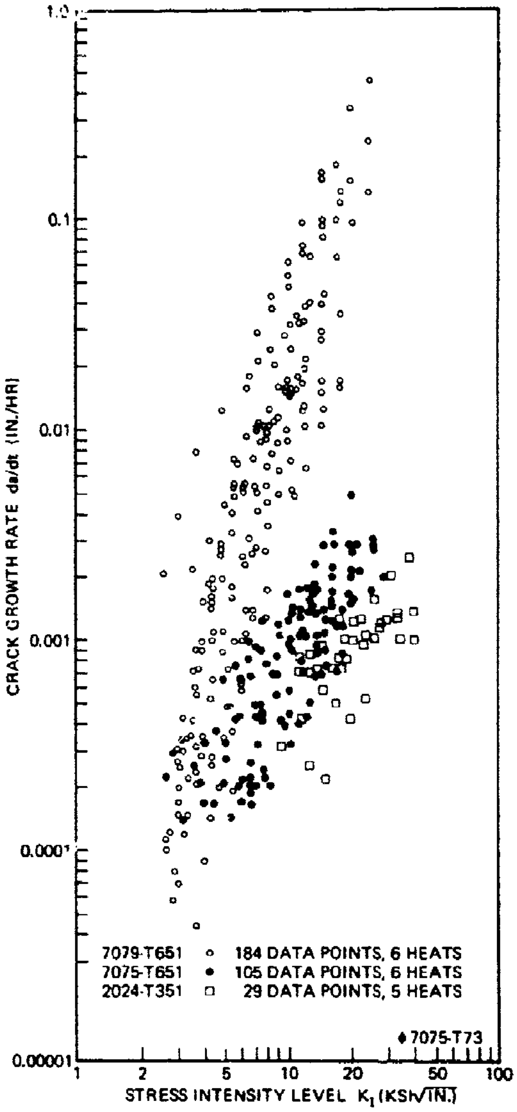


Figure 5. Stress Corrosion Cracking Behavior of Three Commercial Aluminum Alloys Measured Using Precracked DCB Specimens Intermittently Wetted With 3.5% NaCl (6)

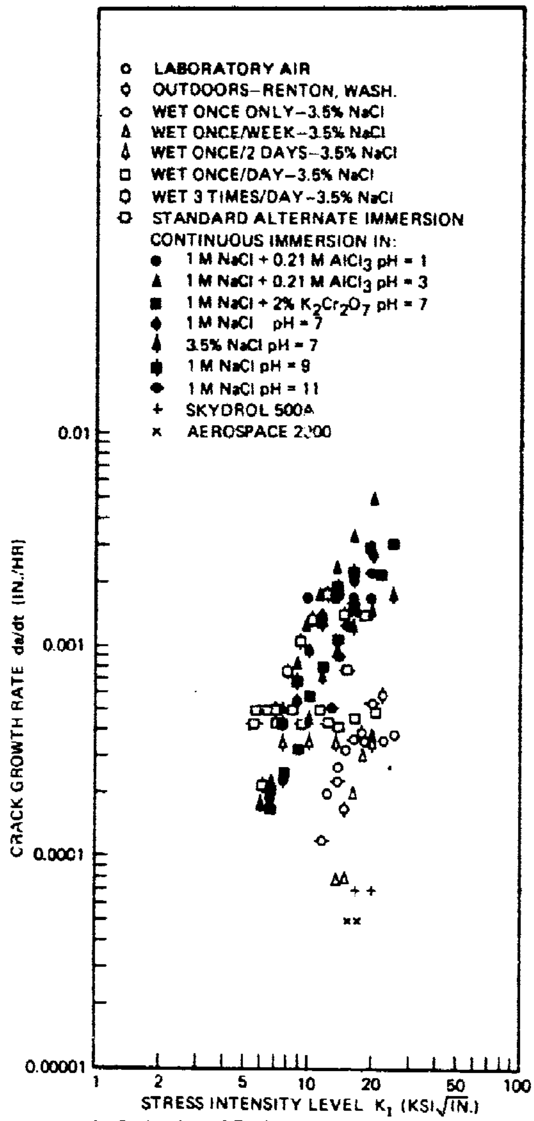


Figure 6. Evaluation of Environments Using the Double Cantilever Beam Specimen for 7075-T651 Aluminum Alloy (6)

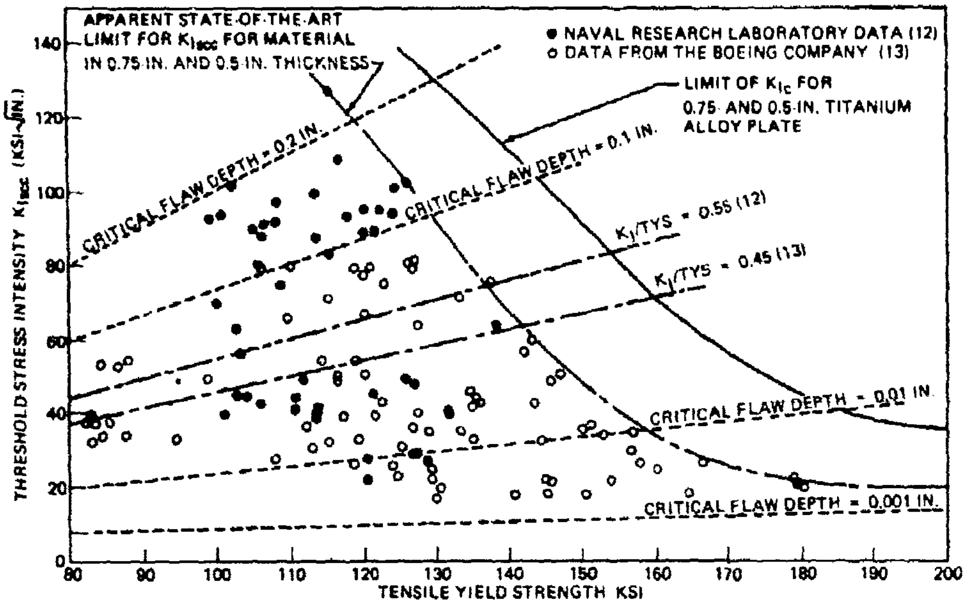


Figure 7. Fracture Mechanics Stress Corrosion Resistance Parameter K_{Iscc} (in 3.5% NaCl) for Various Titanium Alloys Having Range of Yield Strengths

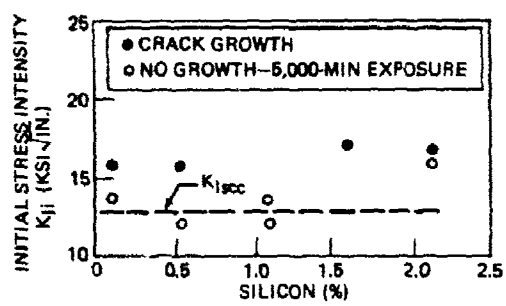


Figure 8. Effect of Silicon on Threshold Stress Intensity K_{Iscc} (280- to 300-ksi Strength Range) (17)

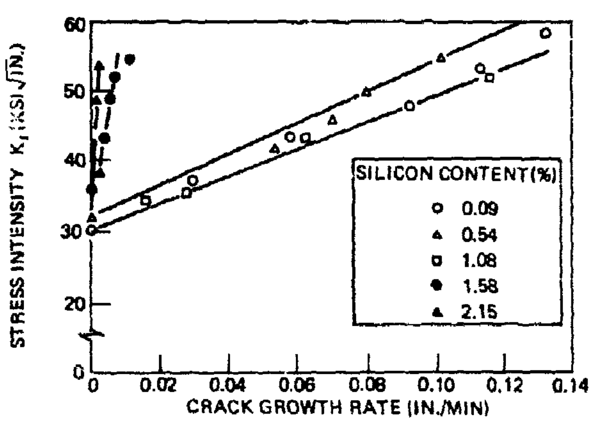


Figure 9. Influence of Silicon Content on Stress Corrosion Crack Growth Rate (280- to 300-ksi Strength Range) (17)

SUSTAINED STRESSES AND THEIR EFFECT ON STRESS CORROSION CRACKING

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1. INTRODUCTION

Many airframe components have failed due to stress corrosion cracking. Most aircraft structures have been made from aluminium alloys and perhaps naturally most failures have been associated with these materials. High tensile stresses may be produced in the material by heat treatment; by bending, machining or fabrication; by poor alignment or interference fits on assembly, and to these may be added sustained stresses arising in service. If these effects are summated stresses approaching the proof strength of the material can be and have been produced.

It would be encouraging if the corrosion part of the process could be inhibited by protective treatments. However, whilst there are obvious advantages with some methods of protection the possibility of corrosive attack will always be present. In the same way the stress component can never be completely removed, hence the stress corrosion properties of all the alloys used must be carefully determined.

With aluminium alloys sustained stress is perhaps the more important component affecting stress corrosion cracking. The direction of the stress relative to the grain direction is particularly important, the "short transverse" grain direction often being the most critical. This paper discusses the origin and magnitude of the stresses which have an important bearing on stress corrosion cracking.

2. RESIDUAL STRESSES IN EXTRUSIONS

2.1. Aluminium Alloys

The solution treatment and quenching of aluminium alloys after extrusion produces a residual stress system with compression on the surface of the extrusion and tension internally. These residual stresses vary with the severity of quench and with the size of the extrusion.

FIGURE 1. compares the residual stress distribution in two extrusions after solution treatment and quenching. The first is from work quoted by Forrest (1) on a 10 inch diameter bar of Al-Cu-Fe-Si alloy. The second extrusion was for an aircraft spar and was approximately $6\frac{1}{2}$ inches by $8\frac{1}{2}$ inches cross section, it was also stretched $\frac{1}{2}$% after solution treatment and quenching.

The residual stress level is not usually a problem in extrusions which can be stress relieved by stretching after solution treatment. Forrest (1) shows that residual stresses may be reduced to a small fraction of the "as solution treated" value by stretching to $\frac{1}{2}$% to 1% permanent stretch. FIGURE 2 shows the effect on surface residual stress of varying amounts of stretch for the aircraft spar extrusion referred to above. The curve shows that the majority of the stress relief possible is achieved with the first $\frac{1}{2}$% to $\frac{3}{4}$% stretch. The stresses

2.1. Contd.

quoted are of course for one ticular extruded section and other more complex shapes may require rather mor tretch to obtain similar stress relief.

Machining operations on str ched extrusions will have the effect of further reducing the residual stre level. Machining should always be undertaken with care however, due to the di ortions which are consequent upon stress relief. Figure 3 shows the effect of machining upon an extrusion containing high residual stresses.

All forming operations on extrusions have the effect of modifying the residual stress distribution. Solution treatment is often required for the forming operation and this solution treatment can also modify the residual stress system. Consequently it is always preferred in these cases to form the extrusion by stretching in order to control the residual stresses. If solution treatment is not followed by stretching very high residual stresses may result. On an aircraft centre section spar cap, solution treated for bending operation, internal tensile residual stresses up to 27 k.s.i. * were found. In this particular case the tensile residual stress was subsequently exposed by drilling of the boom for bolt holes.

If, due to particular circumstance, (for example, the correction of distortions) forming is done on material which is not freshly solution treated very high residual stresses may result. FIGURE 4 gives details of the maximum tensile residual stress produced during four-point bending of a 'T' section extrusion in a variety of heat treatment conditions. If high residual stresses are to be avoided it is most important that all such correction operations should be strictly controlled. An extreme example occurred on an aircraft rib flange of channel section in which tensile residual stresses of 34 k.s.i. were found after correction of relatively small machining distortions.

2.2. Titanium Extrusions

The amount of work which has been done to determine residual stresses in titanium extrusions is somewhat limited. Indications are however that the levels will be low. Extrusions are generally hot stretched at near the annealing temperature which will give a very low residual stress level. For small titanium extrusions where warm drawing is used the residual stresses will be higher but will be compressive on the surface. Such small extrusions will not usually require machining.

Titanium extrusions would of course have the contaminated surface removed by acid pickling and it is then advantageous to follow with an abrasive blasting process to produce favourable compressive surface stresses.

3. RESIDUAL STRESSES IN FORGINGS

3.1. Aluminium Alloys

Residual stresses are generally more of a problem with materials in the forged form than in other forms. The designed shapes and grain directions are often complex giving rise to variable quenching stresses. The quenching conditions with such forgings are also often variable in that steam pockets give locally less severe quenches.

Some aluminium alloys have been found to be more susceptible to stress corrosion in the forged form than in extruded or plate forms. Typical results of stress corrosion tests given by Sprowls and Brown (2) show that, for example, aluminium alloys 2014-T6, 7075-T6 and 7079-T6 have considerably lower stress corrosion thresholds in the forged condition than rolled plate. The same effects have also been noted in U.K. work on alloys of similar composition.

Some early forgings, manufactured in 1945, in Al-Cu-Zn-Mg alloy were cold water quenched and were found to contain residual stresses up to 49 k.s.i. tension. FIGURE 5 shows a failure typical of a forging (shown in FIGURE 6) containing such high residual stresses. Since that time quenching conditions have been improved progressively for those alloys which do not show a marked quench sensitivity. Quenching temperatures have been increased from cold water to water at 85°C, to boiling water and to the step quench at 180°C. These changes have given a progressive reduction in residual stress as the quench rate was retarded, FIGURE 7 gives typical values of residual stress for the various

* Kilopounds per square inch used throughout as unit of stress.

3.1. Contd.

quenching temperatures. With the reduction in residual stress levels, the incidence of failure has been decreased on the forgings susceptible to stress corrosion cracking. The modified heat treatments have also in some instances improved the resistance of the alloy to stress corrosion cracking.

With aluminium alloys such as the 7075-T6 or 2014-T6 types hot quenching above 75°C is not possible due to the reduction in strength properties and other methods of reducing residual stresses must be adopted. Each of the following examples have been found useful in specific cases.

3.1.1. Cold Compression.

This is similar in effect to stretch straightening but it is only possible on forgings of relatively simple shape. On Hiduminium 66 aluminium alloy cold reduction of 5% was found to give 40% reduction in residual stress level as compared with the "as solution treated and quenched" conditions.

Doyle (3) working with aluminium alloy RR58 in the hand forged condition found the amount of cold compression required for stress relief was extremely critical. The application of too much cold compression could result in the production of undesirable tensile residual stresses at the surface.

3.1.2. Special Quenching Techniques.

Barber and Turnbull (4) give an example of the use of water jets to ensure adequate quenching of the internal bore on a large aircraft undercarriage forging. By this means it was possible to eliminate tensile residual stresses at the surface of the bore consequent on a slower rate of quenching at this surface.

3.1.3. Special Quenching Media.

An example of a special quenching medium is the use of cold glycol for quenching, which gives a rate of heat transfer equivalent to a boiling water quench even though the glycol is at cold water temperatures.

3.1.4. Deep Freeze Treatment.

Quenching in liquid nitrogen has been shown to have beneficial effects in reducing residual stresses. The rate of cooling is controlled by a blanket of gaseous nitrogen at the component surface to give the required quenching rate. This process has been used in quenching sheet metal components to avoid distortion.

3.1.5. Shot Peening.

If tensile residual stresses are present in the surface of a forging, due for example to locally machining through the surface compression or to locally retarded quenching, it is possible to modify or eliminate the tension by shot peening. This process produces a layer of compressive residual stress in the surface treated.

3.1.6. Vibratory Stress Relieving.

This process, although still in its infancy has been used with success on large steel fabrications, REFERENCE 5 gives examples of its use on welded structures. It is possible that this technique could prove effective in the stress relief of complex aluminium alloy forgings.

3.2. Steel Forgings.

Residual stresses in steel forgings arise mainly from cooling from forging and hardening temperatures, particularly from quenching during hardening. The tempering treatment will provide an amount of stress relief varying with the tempering temperature and with the material. Low tempers at 220°C give little or no stress relief, high tempers at 400°C to 650°C give more stress relief but this may be offset by the steel being more resistant to tempering as occurs for example with 5% Cr-Mo-V steel.

Bainbridge (6) has shown that with steel forgings the surface residual stresses may be considerably modified by machining and where surface compression is required shot peening is often used.

3.3. Titanium Forgings.

These forgings are usually annealed and may consequently be assumed to be free from residual stress due to forging and heat treatment. However, as with steel forgings, machining may introduce significant surface residual stresses. Again surface stresses may be modified by shot peening.

3.4. Magnesium Alloy Forgings.

Stress corrosion failure is not usually a problem with these forgings since alloys having adequate strength can be chosen which are not susceptible to stress corrosion cracking.

4. RESIDUAL STRESSES IN ALUMINIUM ALLOY PLATE.

The controlled stretching of aluminium alloy plate produces a low level of residual stress. On a typical aircraft component machined from DTD 5020 (2014-T6 type) plate, stretched 1½%, the average level of tensile residual stress was 2 k.s.i. and of compressive residual stress 7 k.s.i. With controlled stretching the plate is adequately and consistently worked and is guaranteed free from serious defects. It is significant that stress corrosion failure in aluminium alloy plate has so far been extremely rare. However, as with aluminium alloys in other forms, it is important that the manufacturing processes, e.g. forming, machining etc, involved in producing airframe components should be controlled so as to avoid the production of excessive residual stresses. In order to facilitate forming and at the same time give a beneficial residual stress distribution shot peening has been used in preference to press bending in the forming of panels (REFERENCE 7).

One problem, encountered during the assembly of an aircraft wing, was to contour the skin to the aerofoil camber and at the same time kink the skin through 6°, the direction of bend being perpendicular to the aerofoil camber. The production method decided upon was to press bend the kink and then to clamp the skin to the rib contour and rivet up the assembly. It was calculated that the maximum surface tension resulting from this procedure was 25 k.s.i. approximately.

In order to avoid stress concentrations in the critical area no rivet holes were drilled at the position of the kink.

5. RESIDUAL STRESSES IN ALUMINIUM ALLOY SHEET.

Stress corrosion failure is rare in aluminium alloy sheet. This is due mainly to two factors, firstly that the corrosion component on commonly used materials is inhibited by a cladding of pure aluminium and secondly, the stress component is often controlled by stretch straightening or stretch forming of the sheet. Clad materials such as DTD 687B (7075-T6) and L.73 (2014-T3) are also extremely resistant to stress corrosion cracking. Samples of these sheet materials have withstood 5 years stress corrosion testing at 100% of the 0.1% proof stress for the material without cracking. This result has been found both on plain specimens and on specimens having drilled rivet holes.

In service many sheet components have tensile residual stresses of half the 0.1% proof stress due to bending to shape and there are no recorded failures. However since assembly and service stresses can be additive to the initial residual stress system it is preferable always to control processes to minimise residual stresses, particularly at sheet edges and hole positions. For this reason stretch forming is always preferred to roll bending.

6. ASSEMBLY STRESSES

So far we have considered the residual stresses which may be produced in individual airframe components during their manufacture. To these residual stresses must be added the "built-in" stresses due to assembly of all the components to form a structure. These assembly stresses may arise due to cumulative tolerances giving interference or considerably loose fits, the stresses being produced either by forcing apart or by clamping together on completion of the assembly. The joining together of sub-assemblies may also give rise to very high assembly stresses in the mating components if they are not exactly in line with each other.

The following are a few examples of the ease with which stresses may be built into an aircraft structure assembly.

6.1. Interference Fits.

On one aluminium alloy undercarriage component the forging tolerances permitted an interference fit between the steel liner and the bore. This produced tensile stresses approaching the proof stress of the material within the forging which caused catastrophic stress corrosion failure in a period of six months.

6.2. Fasteners.

As part of an investigation into the failure of leading edge skin butt straps the stresses in sheet due to rivetting were determined. It was found that tensile stresses up to 16 k.s.i. existed generally in the butt strap due to hammer rivetting, although blind rivetting (similar to CHERRY rivets) produced negligible stresses.

6.3. Contouring by Clamping on Assembly.

FIGURES 8 and 9 show how the butt strap mentioned above was assembled into the leading edge, no forming of the butt strap being done prior to assembly. This method of assembly produced stresses of above 0.1% proof stress on the convex face of the butt strap and stress corrosion failures occurred in service.

6.4. Clamping Up of Loose Fits.

FIGURE 10 shows the assembly of an aircraft fuselage main former. On production it was found that gaps of 0.010 inches were common, these gaps being closed by bolting on assembly. Investigation showed that a tensile stress of approximately 2 k.s.i. per 0.001 inch gap was produced at the outer side of the former on clamping of the joint. In all cases such as this very careful control of design and manufacture is required due to the ease with which catastrophically high assembly stresses may be produced.

6.5. Misalignment.

FIGURE 11 shows the assembly of the spar cap and leading edge skin joint on a medium sized aircraft. Incorrect machining of the spar cap produced a step in the flange across which the butt strap was bent and held by rivetting on assembly. Stresses up to the 0.1% proof stress of the material were produced in the strap during this procedure.

It was fortunate that in this particular case further manipulation of the butt strap during assembly of the leading edge relieved the "built in" stresses to an acceptable value.

7. SERVICE STRESSES

In addition to residual stresses and assembly stresses a third stress component must be considered on most aircraft structural components, that is sustained service stress. Examples of sustained service loads are: for military aircraft, where the aircraft may stand for long periods, undercarriage hydraulic cylinders and crutching components, for civil aircraft having high utilisation, the "1 g" flight case is almost a sustained load.

In considering service stresses the least severe case is the direct stress in simple components, more critical is the stress concentration case arising at holes, radii etc. and in components of complex shape. The worst case occurs when a crack has initiated perhaps by fatigue loading or corrosion or, in new structures, resulting from welding, fabrication or defects. The crack then acts as a nucleus where the stress is considerably increased. The presence of a crack can be of critical importance in a wet environment since the wet stress corrosion properties of some materials, for example high strength steel, are notably poor.

8. CONCLUSION.

As noted earlier in this paper the greatest danger arises when residual, assembly and service stresses combine to produce high tensile stresses. It is the authors' experience that this coincidence of various stresses is often the main cause of stress corrosion failures.

The table FIGURE 12 shows that the total sustained stress may exceed the proof stress of the material. The first objective must always be to improve materials to remove susceptibility to stress corrosion cracking since there will always be some sustained stress present.

It is however always desirable to reduce residual and assembly stresses to the minimum possible by attention to heat treatment, fabrication methods, and assembly techniques. Low stresses will always improve the position as regards fatigue and distortion during machining in addition to the residual stress corrosion hazards.

Acknowledgement

Whilst the various views and opinions expressed in this paper are entirely our own, we would like to express our gratitude to Hawker Siddeley Aviation Ltd., for permission to undertake the paper and to acknowledge the generous assistance given by the company throughout its preparation.

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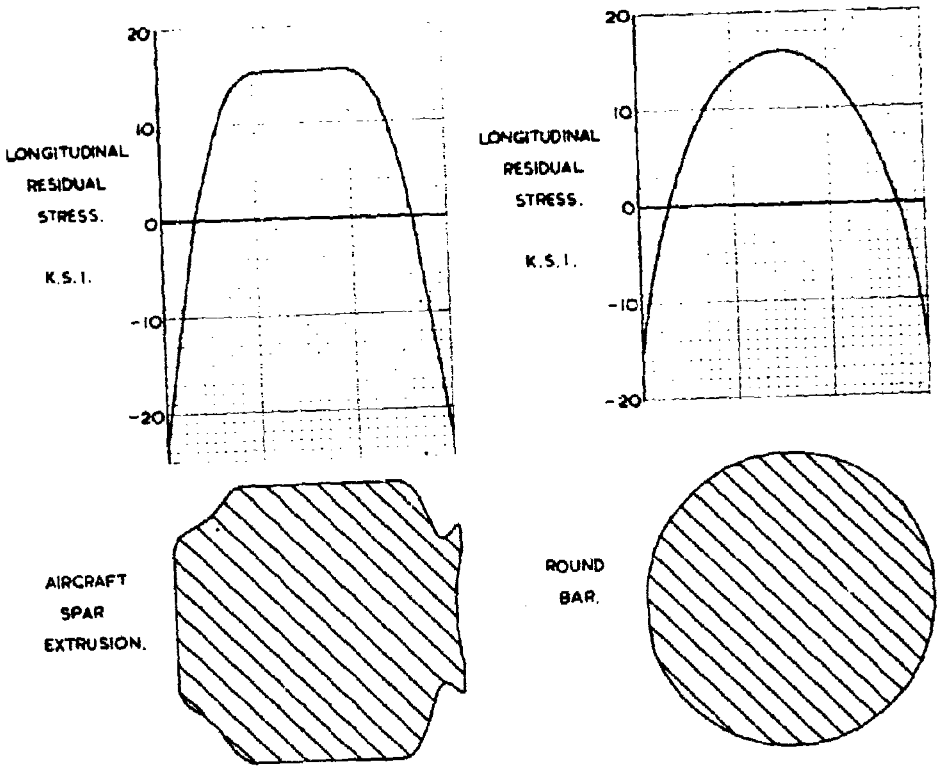

FIGURE 1 Residual Stresses after Solution Treatment and Quenching.

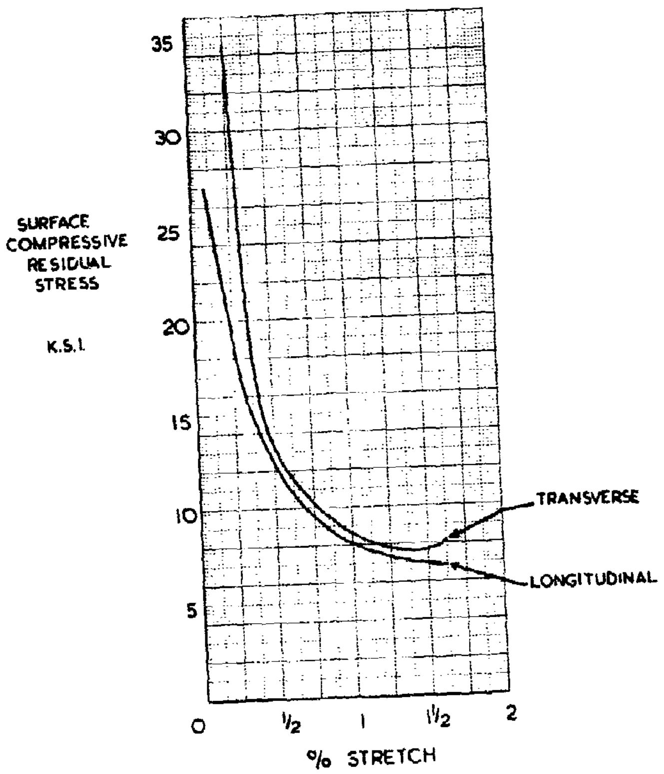

FIGURE 2 Surface Residual Stress against % Stretch - Extrusion 6½" x 8½".

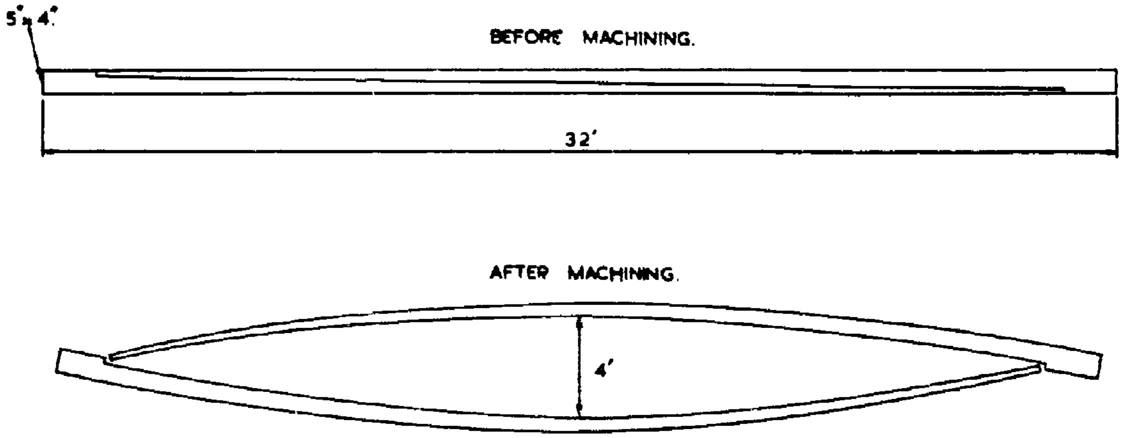

FIGURE 3 L65. Extrusion, Fully Heat Treated but not Stretch Straightened, showing Distortion Due to Machining.

MATERIAL CONDITION ON BENDING.	MAXIMUM TENSILE RESIDUAL STRESS.
FRESHLY SOLUTION TREATED (WITHIN 2 HRS.)	6 K.S.I.
SOLUTION TREATED AND NATURALLY AGED.	15 K.S.I.
SOLUTION TREATED. NATURALLY AGED. BENT AT 135 °C.	15 K.S.I.
FULLY HEAT TREATED.	18 K.S.I.
FULLY HEAT TREATED. BENT AT 135 °C.	17 K.S.I.

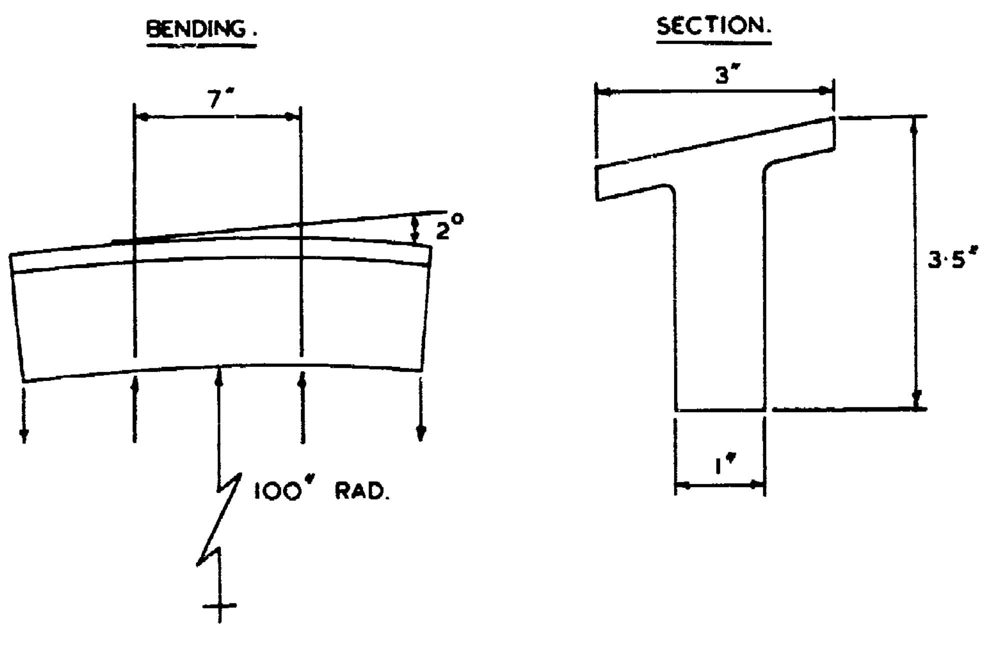

FIGURE 4 Residual Stresses due to Bending of Spar Cap.

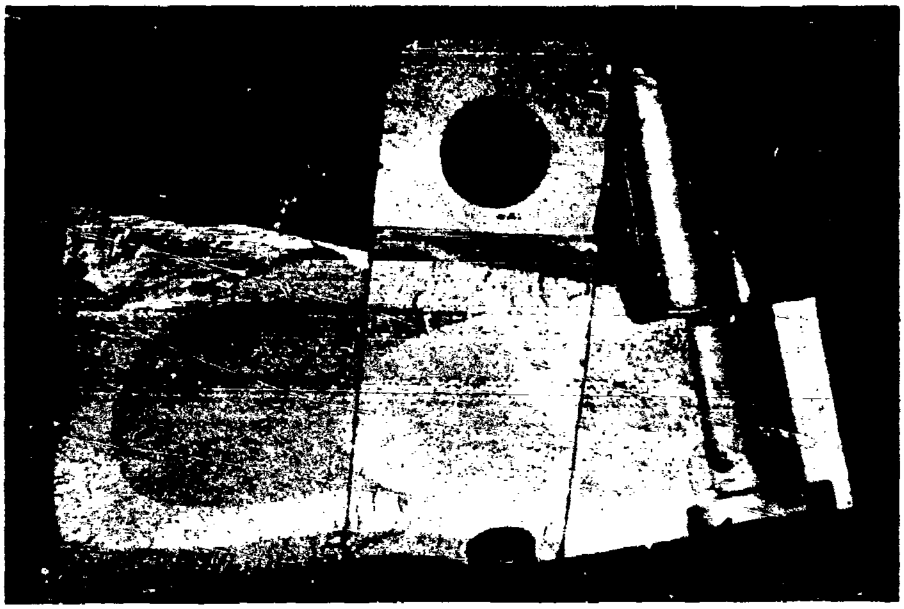

FIGURE 5   Stress Corrosion failure on Undercarriage Beam Forging.

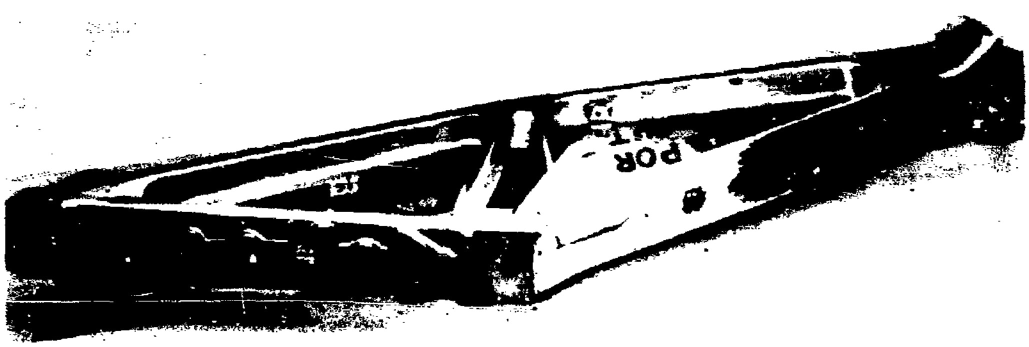

FIGURE 6   Undercarriage Beam Forging.

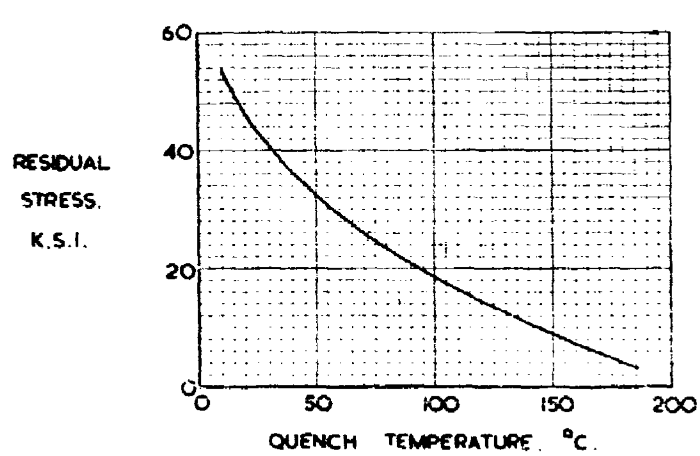

FIGURE 7   Variation of Residual Stress with Quenching Temperature.

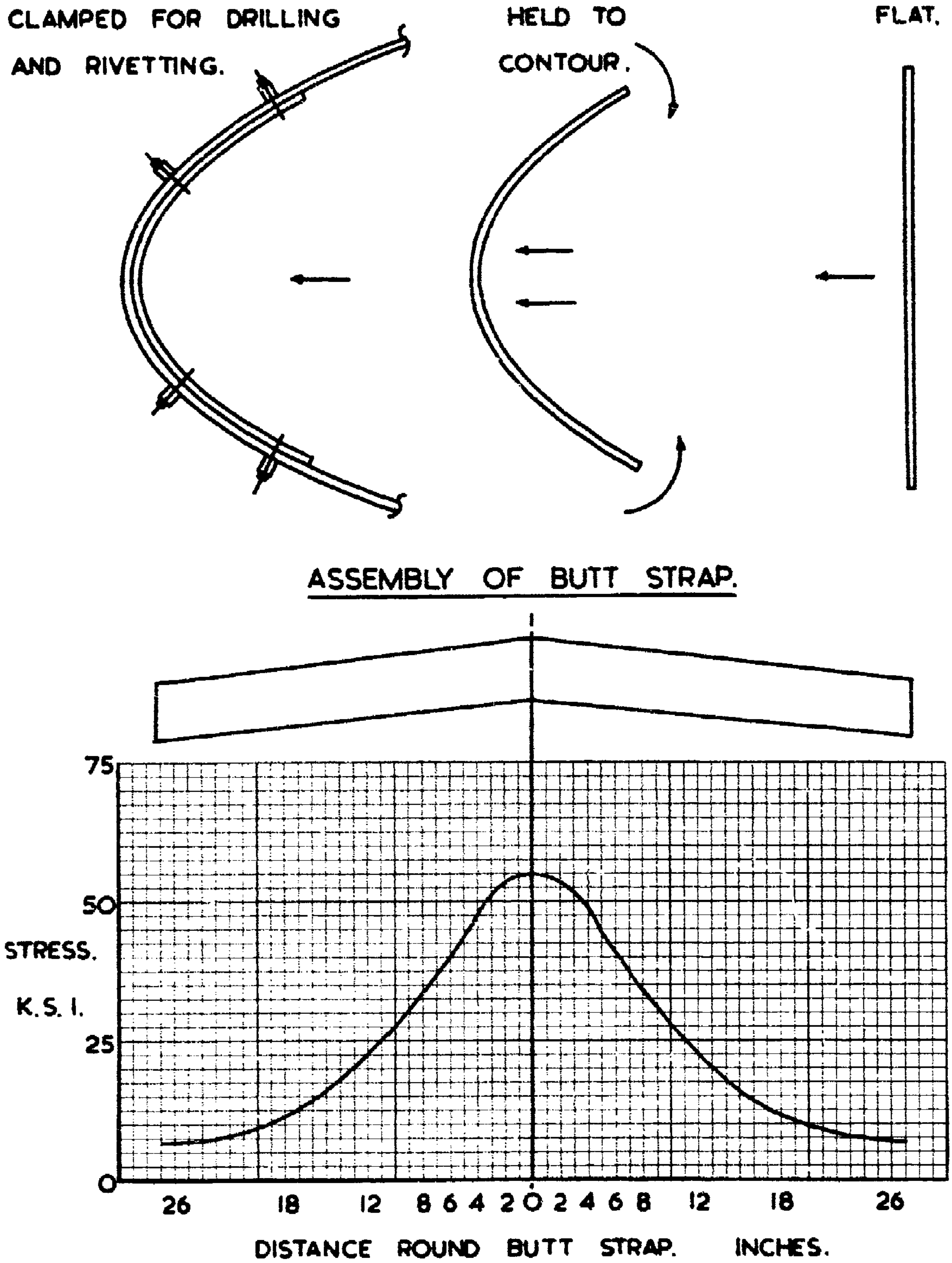

FIGURE 8 Stresses due to Bending to Contour on Assembly.

FIGURE 9 Internal view of aircraft leading edge Showing Butt Strap.

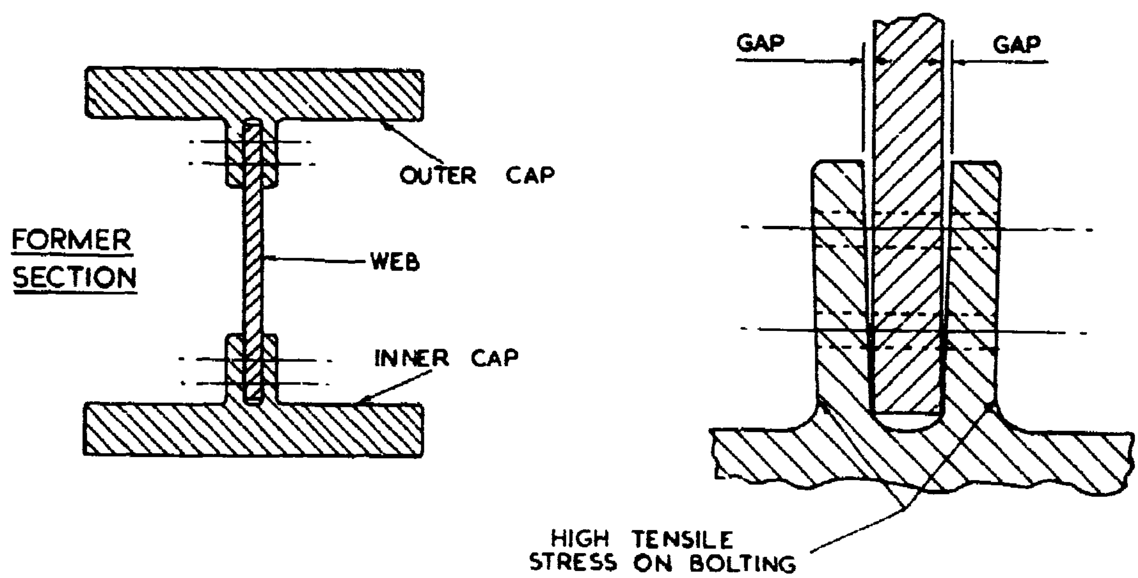

FIGURE 10 Assembly Stresses - Closure of Gaps.

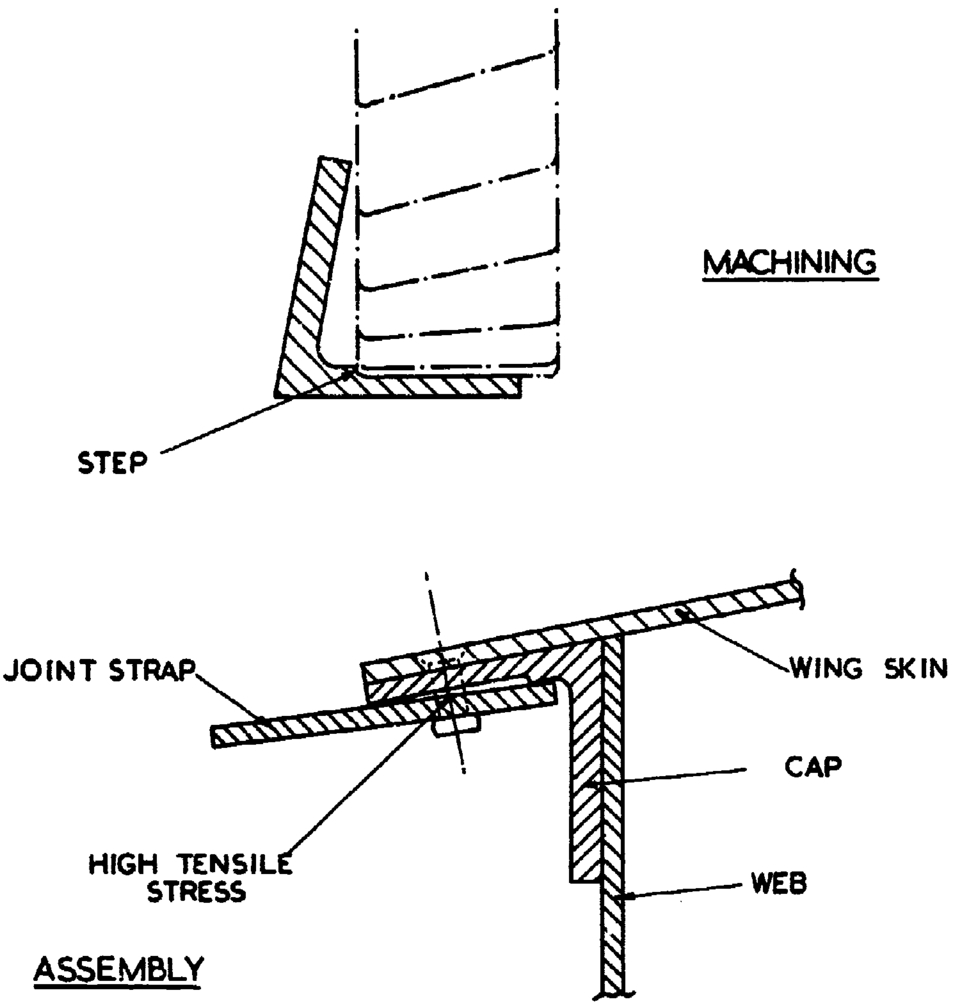

FIGURE 11 Assembly Stresses - Front Spar.

MATERIAL.		MAXIMUM TENSILE STRESS. KILOPOUNDS PER SQUARE INCH.					TOTAL % OF PROOF STRESS.
		RESIDUAL STRESS. HEAT TREATMENT.	MANIPULATION	"NORMAL" ASSEMBLY STRESS.	OPERATIONAL SUSTAINED STRESS INCLUDING STRESS CONCENTRATION	TOTAL POSSIBLE STRESS.	
FORGINGS.	COLD WATER QUENCH.	22-49		11	16-22	67+	100+
	70°C QUENCH.	9-38			16-22	67+	100+
	100°C QUENCH.	9-18		11	16-22	51	76
	STEP QUENCH.	2-4		11	16-22	37	55
EXTRUSIONS.	STRETCHED.	2-6	22-33	11	17	67	100
	RE-SOL. TREATED QUENCHED 85°C & BENT		18-31	11	17	59	88
PLATE.	STRETCHED AND MACHINED.	4	18	22	17	61	91
SHEET.	STRETCH FLATTENED.	4	18	22	17	61	91

FIGURE 12 Effect of Summation of Residual, Assembly and Sustained Service Stresses.

RESIDUAL STRESSES ARISING FROM MACHINING AND FABRICATION

by

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SUMMARY

A brief review is given of how residual stresses may be induced into the surface regions of some metallic materials by machining and fabrication operations. Processes considered include turning, milling, grinding, spark-machining, bending, stretch-forming and welding. An indication is also given of the magnitude and sign of such stresses, which in certain instances can approach the yield strength of a material.

Methods are outlined by which the magnitude of unfavourable residual tensile stresses may be controlled within safe limits or the stresses even completely eliminated, by careful attention to processing parameters and by the use of subsequent thermal or mechanical stress relieving treatments.

RESIDUAL STRESSES ARISING FROM MACHINING AND FABRICATION

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1. INTRODUCTION

Residual stress may be defined as an elastic stress system within a material which does not require the maintenance of external forces for its existence. These stresses must be both balancing tension and compression in a state of static equilibrium. In this paper only macro-residual stresses are to be considered as opposed to the short-range micro- or "tessellated" stresses which may vary in magnitude and sign over distances comparable to the grain size or smaller.

Residual stresses are the consequence of non-uniform plastic deformation and are normally an intrinsic feature of most machining and fabrication operations. Such plastic deformation may be due to mechanical working, thermal expansion and contraction effects, and volume changes arising from microstructural transformations. The magnitude and sign of a residual stress system induced into a material by a given process depends largely upon the processing conditions and the mechanical properties of the material at the time of the induced plastic deformation.

High surface residual tensile stresses can be extremely deleterious as they will act as a tensile mean stress. Where stress corrosion is an important criterion, residual tensile stresses can accelerate stress corrosion of a material susceptible to this form of attack. The presence of a surface residual compressive stress is however normally beneficial, provided that dangerous tensile stresses are not produced elsewhere, and will help to inhibit intergranular attack and reduce the rate of crack propagation. An attempt must therefore be made either to avoid the formation of deleterious stresses or to contain the magnitude of such stresses to acceptable levels during manufacture. This may be facilitated in many instances by careful attention to processing conditions and their control, and furthermore, residual stresses are often amenable to heat treatment and controlled deformation in order to bring about a reduction in their magnitude.

Ideally any stress evaluation should consider the stresses acting in all three principal directions; unfortunately such an approach is not normally a feasible proposition. The stresses acting in one direction only are usually determined and sometimes also those in an orthogonal direction. Surface residual stress evaluations do not always afford a true understanding of the nature of the plastically deformed surface regions. Thus determination of the subsurface stress pattern may be desirable. Stress data are usually obtained by dissection of the worked surface under examination followed by either stress calculations based on changes in physical dimensions or stress measurements by x-ray diffraction or strain gauge techniques.

There is no doubt that a knowledge of the residual stresses incurred during manufacture may be of extreme importance in controlling the final quality of a product and its ensuing service reliability. However, sometimes, even in the field of research insufficient emphasis is placed on the role of residual stresses. The influence of residual stresses on the apparent behaviour of other material properties may be quite significant and so may have to be taken into account.

2. RESIDUAL MACHINING STRESSES

2.1 Turning, Milling, etc.

A number of conventional machining processes are briefly considered. Grinding is however excluded and dealt with separately. These machining processes usually develop compressive stresses in the major working direction of the cutting zone. The magnitude of the stresses is a function of the cutting forces involved and the mechanical properties of the material being machined. Machining variables, which may affect the stress distribution, are cutting tool sharpness, cutting speed, feed rate, depth of cut, tool geometry and cutting fluid. Cutting speeds are normally slower than in grinding operations and material being machined may therefore be kept relatively cool and overheating troubles can be more easily avoided.

An investigation was carried out at Westland Helicopters Ltd. to determine the stresses arising from certain surface working processes. Stresses were measured by the multi-exposure x-ray back-reflection technique[1] on the surface of BS.S99 (2½% Ni-Cr-Mo, 80 tons/in^2) steel test pieces. The test piece designs and direction of stress measurements are given in Figure 1. Process details and results obtained were as follows:-

Process	Machining of test piece	Test Piece Type	Surface Compressive Stress (tonf/in.^{2})
i) Drilling	Centre drilled out with sharp ½in. dia. drill. Section ABC removed.	A	20.1, 18.1
ii) Reaming	Centre drilled out with 31/64in. dia. drill, reamed with ½in. dia. reamer. Section ABC removed.	A	20.0, 16.6
iii) Honing	Drilled and reamed as (ii), hole honed to give 8 micro-inch finish. Section ABC removed.	A	24.5, 29.6
iv) Broaching	½in. dia. hole drilled, then broached to give ½in. square section. Section DEFG removed.	B	36.5, 30.7
v) Band sawing	Cylindrical block cut in half lengthwise.	C	18.1, 18.5 (In cutting direction) 19.1, 19.5 (In feed direction)

Data from the various working processes examined indicate the presence of surface compressive stresses acting in the working direction. Turning can also result in a wide range of compressive stresses, and values in the range 21-39 tonf/in.^{2} have been recorded on BS-S99 steel.

Drilling of blind holes in SAE.4340 (540 HV) steel with a high speed drill which dulled during the drilling operation has been shown[2] to produce untempered martensite on the hole periphery and at the bottom of the hole. Affected material was observed to extend to 0.001in. in depth. High stresses can be developed by such abusive practice. The use of sharp drills, correct drill speeds and feed rates will avoid surface damage. In the same investigation it was found that face milling of SAE.4340 steel produced a thin white surface layer when the wear on the cutting edge of the tool became excessive, whereas with a sharp carbide mill the white layer was almost non-existent. Residual stress patterns for carbide milling of SAE.4340 steel are shown in Figure 2. Tool sharpness is of paramount importance in such machining. A tool with zero wearland produced the minimum stress, whereas the dullest tool, with a 0.016in. wearland, gave the maximum stress. The residual stress in the surface was predominantly compressive although at the extreme surface tensile stresses of increasing magnitude were exhibited as increasingly worn tools were used. Investigation elsewhere[3] has shown somewhat similar effects, and general conclusions were that in milling operations an increase in tool wear, cutting speed, and feed rate induced increased surface tensile stresses. This feature was attributed to a temperature increase at the workpiece surface.

2.2 Grinding

Residual stresses introduced by grinding operations may vary widely in magnitude and sign. The stresses are extremely dependent on the grinding conditions used and are not easily predictable. Variables which will influence the stress pattern are mainly wheel grade, wheel speed, depth of cut, wheel dressing and grinding fluid.

During grinding, extreme heat can be generated at the workpiece surface in contact with the grinding wheel. The heated region tries to expand but is restrained by the surrounding cold metal and thus becomes compressively stressed, resulting in plastic deformation. As the momentary contact between the grinding wheel and the workpiece terminates, the previously heated region is suddenly cooled by heat dissipation to the surrounding material, and contraction occurs producing surface tensile stresses depending on the grinding conditions applied. The main problem in grinding operations therefore, is to avoid burning of the extreme surface. Severe heating of ferrous materials can be sufficient to form a skin of austenite which subsequently transforms to martensite due to the steep temperature gradients and rapid cooling rates involved. The untempered martensite formed will be compressively stressed but is inherently brittle and subject to cracking. Below the martensitic layer any overtempered structure will exhibit tensile stresses. These residual stresses develop as a consequence of the expansion which takes place during the austenite to martensite transformation and the contraction associated with the tempering of martensite.

Abusive grinding, especially of low alloy steels, may result in either overtempering, or both rehardening and overtempering burns. Such phenomena may be revealed by etching the ground components in a suitable acid solution. Where good grinding practice is used, no evidence of such deleterious features should be present on inspection of ground component surfaces. Ideally, minimum surface heating should occur, and the material should be compressively stressed due to plastic deformation and work hardening of the surface fibres.

The residually stressed surface layers often extend to a depth of 0.010in. or more. Steep stress gradients and changes in sign can occur over the first few thousandths of an inch immediately below the surface. The peak stress is usually located at a subsurface position. It has been shown for high strength steels[2], that conventional grinding wheel speeds of 6000 ft/min tend to produce higher tensile stresses (Fig.3) compared with slower speeds of 4000 and 2000 ft/min. The depth of affected material is also reduced with a decrease in wheel speed. Use of wheels of increasing hardness increases tensile stresses and the depth of stressed material (Fig.4). The greater the down feed the higher is the tensile stress and depth of material affected (Fig.5). Refined grinding procedures ("low stress" conditions[2]) resulted in affected depths of about 0.001in. only with low surface compressive stress. Further refined grinding using "low stress" conditions, a low grinding speed and a soft wheel gave a shallow surface compression with an extremely low subsurface tensile stress (Fig.6).

In an investigation at Westland Helicopters Ltd. of a major component failure, thick coupon specimens of BS.S28 (4½% Ni-Cr-Mo, 100 tonf/in.^{2}) steel were deliberately ground abusively to produce grinding burns, and gently for comparison purposes. Nital etch inspection revealed the presence of rehardening and overtempering burns. The subsurface stress distribution was determined, using the x-ray back-reflection technique, for regions showing overtempering burns, rehardening and overtempering burns, and no apparent surface defects. The results of residual stress data from this investigation are given in Figure 7. The stress distribution/depth curve for the defect free area, considered to be representative of satisfactory grinding practice, shows a compressive stress of 32 tonf/in.^{2} at the surface. The material was affected to a depth of about 0.0015in. with a moderate subsurface peak tensile stress. The compressive surface stress was probably due to work hardening effects, although as only 0.010in. of metal was removed by grinding the initial heat treatment stress system may have exerted some influence. The form of the stress distribution curve indicated that heating effects during grinding were minimal. The overtempered ground region exhibited a surface tensile stress of 8 tonf/in.^{2} with a peak stress of 23 tonf/in.^{2} at approximately 0.001in. below the surface. The material was affected to a depth of 0.003in. The tensile stresses were developed due to overheating of the extreme surface together with the formation of overtempered martensite. The severely burnt region of rehardening and overtempering showed a compressive stress of 19 tonf/in.^{2} at the surface with a fairly steep stress gradient giving a maximum tensile stress of 29 tonf/in.^{2} subsurface at a depth of approximately 0.0015in. The total depth of affected material was over 0.010in. Considerable overheating of the original structure must have taken place to have given such a large area of tensile residual stress. The surface compressive stress could have resulted from the formation of martensite near the surface. Regardless of the grinding conditions imposed the peak stress always occurred at a depth not more than 0.0015in. below the surface.

Grinding of titanium alloys can present certain problems; however, it has been shown[4] that such material can be ground, with precautions and at lower than normal speeds, to give low residual stresses. Moderately light down feeds must be used and frequent wheel dressing carried out to obviate wheel loading. The loading problem is largely due to the low thermal conductivity of titanium leading to high surface temperatures, and also to the readiness with which the surface galls. Due to the strong affinity of titanium for oxygen, the metal surface will be prone to atmospheric contamination during grinding if extremely high temperatures are generated. Absorption of oxygen at the surface can result in the formation of a thin oxygen rich layer of stabilised alpha titanium, which is hard and brittle. The microstructure at the surface of satisfactorily ground Hylite 51 titanium alloy (4Al-4Sn-4 Mo- 0.5Si, 90 tonf/in.^{2}) is shown in Figure 8, and a brittle outer skin formed by abusive grinding at the extreme surface of the same alloy is shown in Figure 9. Where a choice of machining is possible for titanium and its alloys, it is advantageous to use an alternative process, such as turning or milling.

To ensure satisfactory grinding quality the actual conditions employed may require experimental determination. Satisfactory grinding parameters, once determined, must be strictly adhered to. The application of the more sophisticated grinding techniques may of course be economically justifiable only for vital highly stressed components.

2.5 Spark-Machining

Spark-machining (or EDM) is a process which can be applied to any electrically conductive material. Precise control of the various spark-machining parameters is however of extreme importance in order to achieve desired metal removal rates or to minimise the extent of workpiece surface damage which always occurs.

Spark-machining causes intense local heating of the material being machined in the immediate vicinity of the electrode. The shallow saucer-shaped pits produced on an eroded surface are usually associated with a layer of re-solidified metal and residually stressed surface regions. Cracking can be exhibited by the damaged surfaces of all spark-machined materials. Surface damage in spark-machined NCMV (Ni-Cr-Mo-V, 120 tonf/in.^{2}) steel is shown in Figure 10.

It has been shown[5] that for various annealed materials, surface tensile stresses of an appreciable magnitude are developed. These stresses are confined to the immediate surface regions and show a steep stress gradient before giving a low compressive stress at about 0.004in. below the surface (Fig.11). Tensile stresses in heat treated steels have been reported[6] to give average values of 34 tonf/in.^{2} in a carbon steel (510 HV) and 67 tonf/in.^{2} in a high speed steel (800 HV). The thickness of the stressed layer was 0.0007-0.001in.

The high tensile stresses and surface damage inherent in the spark-machining process must be eliminated in order to avoid impairing material properties. Affected surface layers must therefore be removed. This can be facilitated in many instances by chemical milling techniques or a machining treatment such as honing. Determination of the effective depth of a damaged surface can however present some problem as its extent may not

necessarily be accurately ascertained by microscopic means. Damage assessment on the basis of fatigue performance for PRN-2 (C-Si-Mn-Cr, 595 HV) steel showed that 0.030in. had to be removed to restore the original endurance limit[6]. This improvement was attributed largely to the elimination of microcracks. The depth of material to be removed may be minimised by the use of a controlled multi-pass technique consisting of perhaps roughing and intermediate passes of the electrode followed ideally by a final pass to give optimum surface finish. A thin layer only of affected material should be introduced by such a procedure. However, it has been shown[7] that even at low machining rates occasional severe local damage can occur.

3. RESIDUAL FABRICATION STRESSES

3.1 Forming and Stretch-Forming

3.1.1 Forming may be considered to include any type of work applied in fabrication processes such as bending, flanging, twisting, dimpling and rectification where a load is applied to produce permanent deformation of the material. A permanent set in the material produced by plastic deformation results in the formation of residual stress as illustrated in Figure 12 by the bending of a simple beam. When a beam is bent beyond the elastic limit and is still under the applied bending load the outer fibres will have yielded plastically on both tension and compression faces. On releasing the restraint, elastic springback occurs and the final stress distribution through the beam is a compressive stress on the tension face and a tensile stress on the compression face. The degree of such residual stresses is dependent on the degree of deformation, springback and the mechanical properties of the material at the time of deformation.

It is advisable in many instances to limit the magnitude of residual stresses when aluminium alloy extrusions such as BS.L64 (Al-Cu-Mg-Si-Mn, 22 tonf/in.^{2}), BS.L65 (Al-Cu-Mg-Si-Mn, 28 tonf/in.^{2}), DTD.5044 (Al-Zn-Mg-Cu-Mn, 32 tonf/in.^{2}) and BS.L88 (Clad Al-Zn-Mg-Cu-Mn, 30 tonf/in.^{2}) sheet material are used in aircraft manufacture. A tensile stress limit of 4.5 tonf/in.^{2} has been considered necessary for vital components. In certain circumstances however, where failure due to stress corrosion would not be disastrous, the tensile stress limit may be raised to 9 tonf/in.^{2} if the lower limit is not obtainable by any normal manufacturing procedure.

Considerable work[8] has been carried out at Westland Helicopters Ltd. on a whole range of high strength aluminium alloys to correlate surface residual stresses with surface residual strains (i.e. permanent elongation of surface fibres) resulting from the forming of these alloys, under conditions of four point loading, when carried out cold, in the solution treated condition, in the fully heat treated condition and, for some alloys, after controlled stretching at room and elevated temperatures. Thus it is possible to determine the maximum residual strain to which these alloys may be formed, when in a given condition, such that the residual tensile stress limit is not exceeded.

Permitted amounts of deformation are expressed as percentages of the outer fibre strain for extruded bar and plate, and for sheet material as minimum bend radii consistent with tolerable residual stress levels. The amount of deformation permitted varies with the condition of the material. Comparatively low strength material in the softened or annealed condition is amenable to large deformations, whilst only small deformations are permitted for materials which have been strengthened, for example by heat treatment.

Typical results of such investigational work, where residual stress data were determined by the x-ray back-reflection technique, are given in Table I which indicates the degree of strain corresponding to tensile stress limits of 4.5 or 9 tonf/in.^{2}. The graphs in Figure 13 indicate the magnitude of residual stress variation corresponding to small changes in residual strain for DTD.5044 aluminium alloy extruded bar formed in various conditions.

Thus where it is required to keep forming stresses to a minimum, it is necessary to form the material in its low property condition and, where possible, to use precipitation as a partial stress relieving treatment. Reference to such residual stress/residual strain graphs enables a reliable estimate to be made of the stresses resulting from forming operations.

3.1.2 Stretch-forming is a fabrication technique widely used in the aircraft industry for the shaping of sheet material, especially in aluminium alloys. The process involves bending or stretching in such a manner that plastic deformation occurring is primarily in tension. Any ductile material is normally suitable for this forming operation.

The principle of the process can be explained by reference to a simple beam of uniform cross section[9], although where compound curvatures are involved the resultant stresses and strains may be extremely complex. As already discussed in 3.1 if a beam is bent to a given radius, such that the outer fibres are plastically deformed, on release of the applied load, springback will occur and the stress sign will be reversed. In stretch-forming however, (Fig.14), a tensile load is applied to the ends of the materials during bending across a rigidly supported forming-die. Thus the longitudinal axis is stretched as the beam is bent. The tensile stress due to bending is increased and the compressive stress on the concave surface decreased. As the applied load is further increased the compressive stress present is eventually eliminated and the whole cross section will become stressed in tension with the maximum stress in the outer fibres. When the stress in the outer fibres exceeds the elastic limit of the material, the strain will proceed with little increase in stress. Eventually the stress will become reasonably uniform across the section in the plastic range. On removal of the applied tensile load the material will contract fairly uniformly around the contour of the forming-die.

From the residual stress aspect the stretch-forming process can be used to considerable advantage. Due to the relatively uniform stress gradients developed, resultant stresses are reduced to low levels and springback becomes less of a problem in fabrication. If stretched sufficiently during forming, residual stresses for aluminium alloys may become negligible[10].

3.2 Residual Welding Stresses

Welding operations cause severe temperature gradients which result in residual stresses when the weldment has cooled to room temperature. These stresses are unavoidably incurred as a consequence of restraints which prevent dimensional changes taking place freely on cooling from the welding temperature. The magnitude of the stresses depends upon such factors as the workpiece material, size and shape, welding process and the welding technique. The residual stress patterns developed due to welding processes may be extremely complex.

Stresses may be conveniently classified[11] as local residual stresses and reaction residual stresses.

i) Local residual stresses arise due to expansion and contraction of the weld metal and heat affected zone, to volume changes due to metallurgical transformations, and to plastic deformation of adjacent regions of the weld and parent material at elevated temperatures. The weld metal and heat affected zone undergo significant contraction stresses on cooling down from high to ambient temperatures. At room temperature this weld contraction may result in significant residual tensile stress levels in the weld metal and parent material adjacent to the fused zone. High heat inputs from pre-heat and the welding process itself give slow cooling rates, whereas welding of thick sections without additional thermal treatment may result in very fast cooling rates equivalent to that of a fairly rapid quench.

These local residual stresses act in three dimensions and occur whether the parent material is allowed to move freely or is restrained. They are only exhibited in the vicinity of the weld and generally decrease rapidly in magnitude with increasing distance from the weld deposit.

Tests conducted by many investigators have shown that local residual tensile stresses approaching the yield strength of a material may be produced as a result of welding. A typical residual stress distribution for a butt-welded joint in the as-welded condition is illustrated in Figure 15.

ii) Reaction residual stresses arise due to restraint applied to the parent material during the welding process, such as in the welding of a rigid assembly. These stresses are generally uniform and affect a welded structure over considerable distances in contrast to local residual stresses.

4. RELIEF OF RESIDUAL STRESSES

4.1 Thermal Treatment

Elimination or reduction of residual stresses may be accomplished by stress relieving heat treatment after machining, forming or welding.

In machining operations stress relief will prevent the distortion that would be caused by asymmetric metal removal and release of residual stresses. Stress relief of fully heat treated steels can be achieved by re-tempering just below the original tempering temperature. Generally, stress relief at about 500°C is sufficient for a range of steels such as BS.S61 (12% Cr. 55 tonf/in.^{2}) and BS.S99. Low temperature stress relief may significantly lower residual grinding stresses. Treatment at 200°C has been found to reduce the susceptibility of satisfactorily ground and overtempered BS.S28 steel surfaces to cracking during subsequent chemical processing treatment.

Thermal treatment will not completely relieve high residual stresses in aluminium alloys. The temperature requirements necessary for significant stress relief are higher than the artificial ageing temperatures. However, partial stress relief may occur during ageing.

Pre- and post-weld heat treatments can restrict the formation of residual welding stresses. Stress relieving treatments applied to low alloy steels at about 650°C, followed by slow cooling to ensure that no additional stresses are incurred, improve the metallurgical structure. Prolonged heating at somewhat lower temperatures may also effect some moderate reduction in the residual stress level. In an investigation into the tungsten-inert-gas welding of titanium alloy sheet[12], residual tensile stresses of 27 tonf/in.^{2} in Ti-6Al-4V material were completely eliminated by a stress relieving treatment at 790°C for 15 minutes.

4.2 Mechanical Treatments

4.2.1 Controlled-stretching may be employed to reduce stresses, particularly in quenched aluminium alloy sheet and uniform extruded sections. Stresses are relieved by stretching in the plastic range. The material is strained in tension and those regions of a section already stressed in tension attain the plastic state first and continue to strain with little increase in stress (Fig.16). Straining is continued until the whole cross section is in the plastic state. When the straining load is released the magnitude of the compressive and tensile stresses across the section will be relatively small.

Stress relief may also be effected in compression.

4.2.2 Reversed bending can be applied to reduce either the level of surface stress or even to reverse the sign of the original stress distribution (Fig. 17). This may be accomplished by an initial slight over-bending followed by a reverse corrective bending. However, a complex stress pattern may result within the bulk of the material.

4.2.3 Harmful machining stresses may warrant removal by a secondary surface working process such as honing, abrasive tumbling or shot-peening. Peening may also be employed to eliminate surface tensile stresses in weldments and impart a compressive stress system. Work[13] on AISI.52100 (1%C-Cr) steel has demonstrated that abrasive treatment with aluminium oxide chips after grinding eliminated a surface tensile stress and introduced a high stress of opposite sign with low subsurface tensile stress (Fig. 18).

4.2.4 Welded structures such as pressure vessels may be stress relieved by hydrostatically pressurising the vessel beyond the yield point. When such loading is released the magnitude of the residual stress will be considerably reduced. Plastic deformation of this kind must be applied well above the transition temperature of the material.

5. STABILITY OF BENEFICIAL RESIDUAL STRESSES

Compressive stresses are considered to be extremely favourable from a stress corrosion aspect. Stress corrosion degradation of aircraft components may be prevented[14] by the presence of surface compression. Compressive stresses, as previously discussed, may be induced by machining operations; however, a more predictable compressive stress distribution, extending to some appreciable depth below the surface, may be achieved by cold working the surface by such methods as shot peening, surface rolling and ballising. Shot peening may also be used for the rectification of severely distorted components and for controlled forming operations. However, the formation of a compressive stress system accompanied by the formation of an exposed balancing tensile stress of a significant order would be most undesirable. Also, where the surface compression is extremely shallow, with perhaps attendant steep subsurface stress gradients, removal of the compressively stressed surface layers may lead to exposure of deleterious tensile stresses. Chemical treatments or even chemical inspection techniques may be sufficiently aggressive to remove a thin layer of compressively stressed material.

It follows, therefore, that where compressive surface stresses are present their stability may be of extreme importance. The continuing effectiveness of a favourable residual stress state during the life of a component depends upon the stability of such a stress system irrespective of changes in loading conditions or environment. Any "fading" of the original residual compressive stress could allow a harmful service stress to predominate.

As part of a larger investigation[15], work was carried out at Westland Helicopters Ltd. to establish to what extent the compressive stresses imparted by shot peening were maintained at moderately elevated temperatures. Tests were carried out on BS.S99 steel and 18% Ni maraging steel (grade: 110 tonf/in.^{2} 0.2% P.S.) to determine the effect of thermal treatments, at up to 500°C, and cyclic stressing, at up to 300°C, on the stress stability of turned and peened surfaces. X-ray stress measurements showed that in certain instances some reduction in surface compression due to thermal treatment occurred (Table II). However, in general, the subsurface stress distribution was little affected (Fig. 19). A combination of fatigue loading and prolonged thermal soaking by cyclic stressing at 300°C near the endurance limit reduced the magnitude of residual compression of both the surface and peak subsurface stress. Vacuum annealing at 500°C gave a surface compressive stress of 4.5 tonf/in.^{2}, nevertheless considerable subsurface compression remained.

The degree of stress stability exhibited by the two steels examined may not necessarily be shown by these and other materials given less intense surface cold working treatments, and exposed to various environments. Thus, attention to residual stress stability may become of increasing importance as operational conditions of vital structures continue to become more demanding.

6. CONCLUSION

Machining and fabrication operations may bring about significant changes in the residual stress pattern of the surface and subsurface regions of a material. If these stresses are tensile, and of an appreciable magnitude, they may exert a deleterious influence on stress corrosion resistance.

Uncontrolled manufacturing procedures may permit abusive practices to occur resulting in undesirable stresses. Where considered justifiable, processing parameters should be precisely defined and their implementation carefully controlled to ensure that dangerous tensile stresses are completely avoided or at least reduced in magnitude to acceptable levels.

Care and attention to manufacturing detail may be amply rewarded by components and structures giving reliably safe service life and the avoidance of catastrophic failures.

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TABLE 1

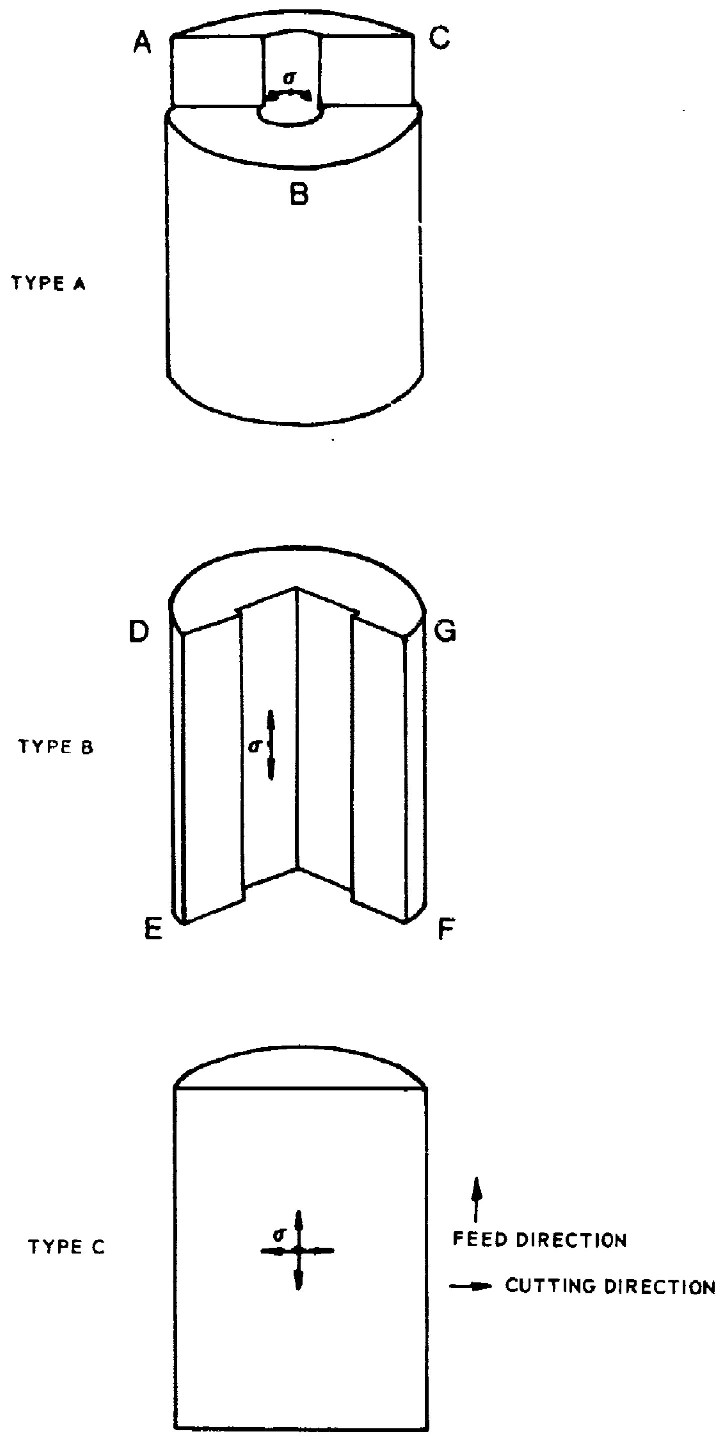
Typical Residual Strain Limits Determined for High Strength Aluminium Alloys

Material	*Heat Treatment and Forming Sequence*	*% Residual Strain*	
		4½ tonf/in.² residual tensile stress limit	*9 tonf/in.² residual tensile stress limit*
DTD. 5044 Extruded bar (Al-Zn-Mg-Cu-Mn, 32 tonf/in.²)	Solution treated (85° quench). Cold formed (a) 2 hrs, (b) 6 hrs, after solution treatment. Precipitated.	a) 0.2 b) 0.2	> 3.0 > 3.0
	Solution treated (85°C quench). Precipitated. a) Cold formed. b) Hot formed (140°C)	a) 0.2 b) 0.4	0.6 > 1.0
	Solution treated (cold quench). Controlled stretched within 4 hrs of quenching. Precipitated. Cold formed.	0.1	0.3
	Solution treated (cold quench). Controlled stretched within 4 hrs of quenching. Cold formed (a) 24 hrs, (b) 1 month after solution treatment. Precipitated.	a) 0.1 b) 0.1	1.0 0.6
BS. L65 Extruded Bar (Al-Cu-Mg-Si-Mn, 28 tonf/in.²)	Solution treated (cold quench). Cold formed (a) 2 hrs, (b) 6 hrs after solution treatment. Precipitated.	a) 0.8 b) >1.0	> 4.0 > 4.0
	Solution treated (cold quench). Precipitated. Cold formed.	0.2	0.7
	Solution treated, (cold quench). Controlled stretched (1.5%) within 4 hrs of quenching. Precipitated. Cold formed.	0.2	> 1.0
	Solution treated, (cold quench). Controlled stretched (1.5%) within 4 hrs of quenching. Cold formed (a) 24 hrs, (b) 1 month after solution treatment. Precipitated.	a) 0.2 b) 0.2	> 1.0 > 1.0

TABLE II

X-Ray Surface Stress Measurements on Turned and Shot Peened Surfaces Before and After Thermal Treatment

Material	*Condition*	*Residual Compressive Stress tonf/in.$^{-2}$ (±5)*
BS.S99 (2½% Ni-Cr-Mo, 80 tonf/in.^{2})	Turned	33.5
	Turned + 200°C/24 hours	33.0
	Turned + 200°C/24 hours + 300°C/120 hours	33.0
	Shot peened	30.0
	Shot peened + 200°C/24 hours	24.0
	Shot peened + 200°C/24 hours + 300°C/120 hours	15.5 (±3)
18% Ni Maraging steel (grade: 110 tonf/in.^{2}, 0.2% P.S.)	Turned	20.5
	Turned + maraged (475°C/8 hours)	2.5 (±3)
	Maraged + shot peened	33.5
	Maraged + shot peened + 200°C/ 24 hours	28.0
	Maraged + shot peened + 200°C/ 24 hours + 300°C/120 hours	20.0



σ : DENOTES POSITIONS AT WHICH STRESS
MEASUREMENTS WERE MADE

Fig. 1 Machining test piece designs

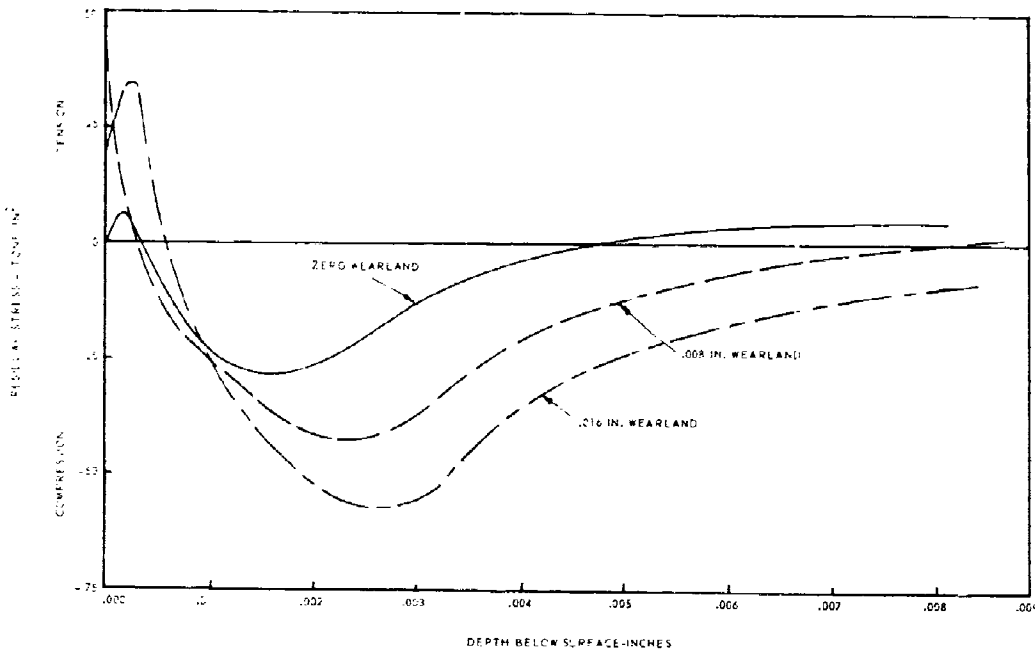


Fig. 2 Residual stresses in milled surface of SAE.4340 steel (540 HV). (Field and Kahles[5])

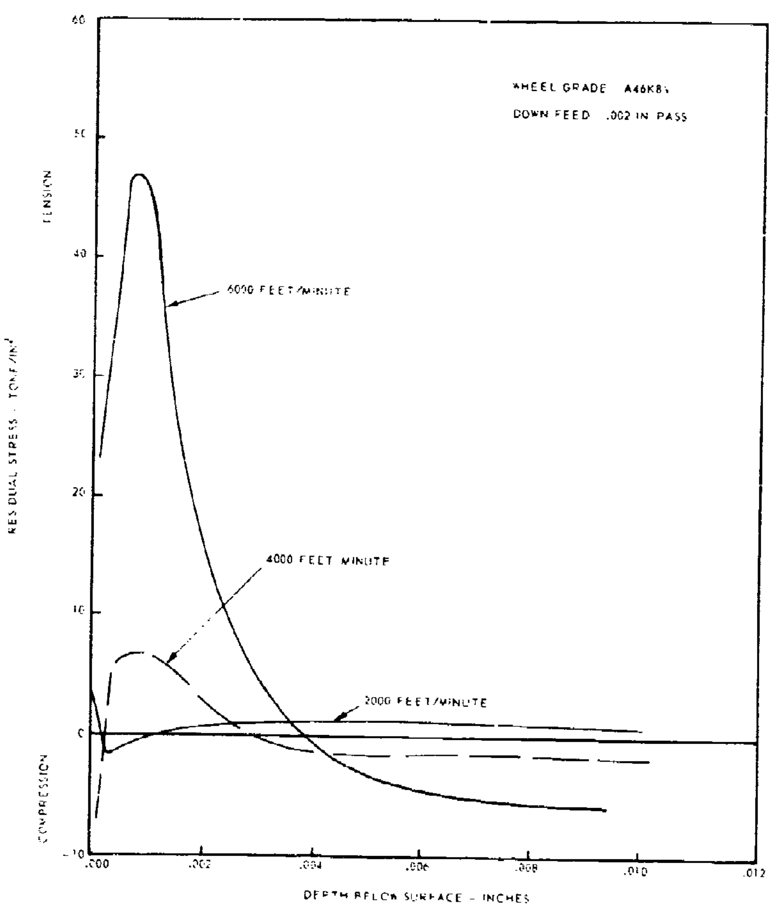


Fig. 3 Effect of wheel speed on residual grinding stresses of SAE.4340 steel (540 HV). (Field and Kahles)

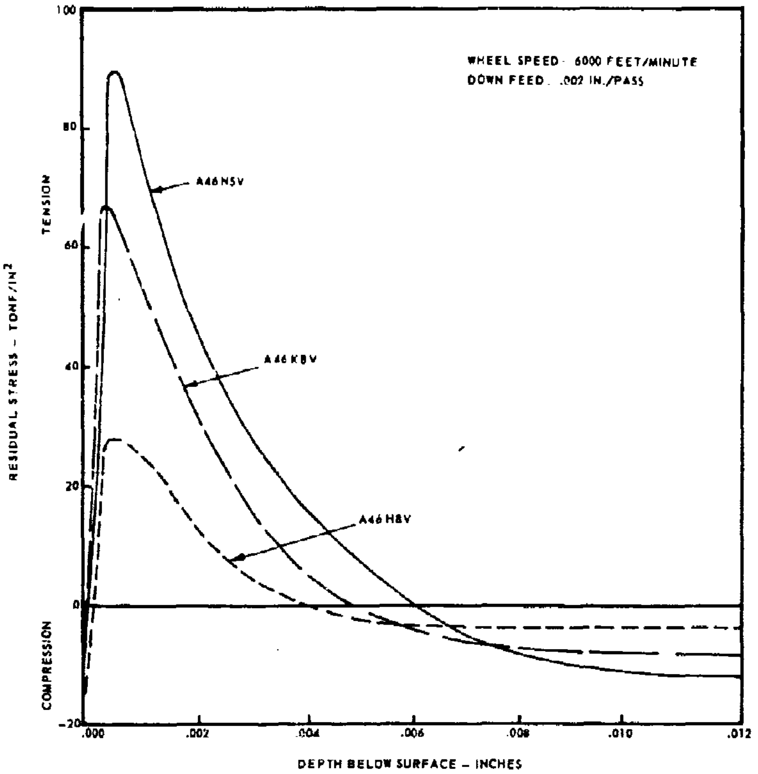


Fig. 4 Effect of wheel grade on residual grinding stresses of D6AC steel (610 HV). (Field and Kahles[2])

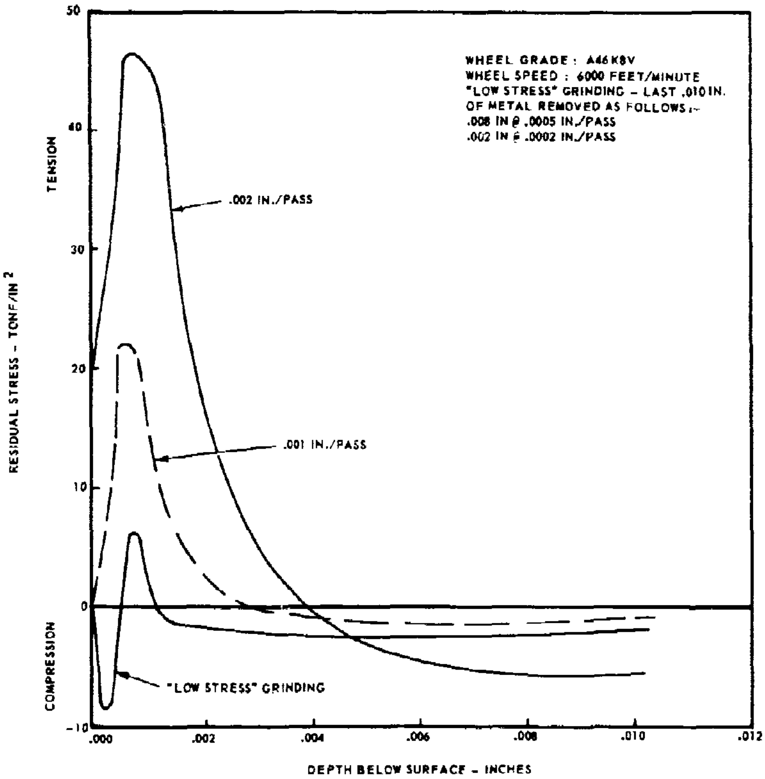


Fig. 5 Effect of down feed on residual grinding stresses of SAE 4340 steel (540 HV). (Field and Kahles[2])

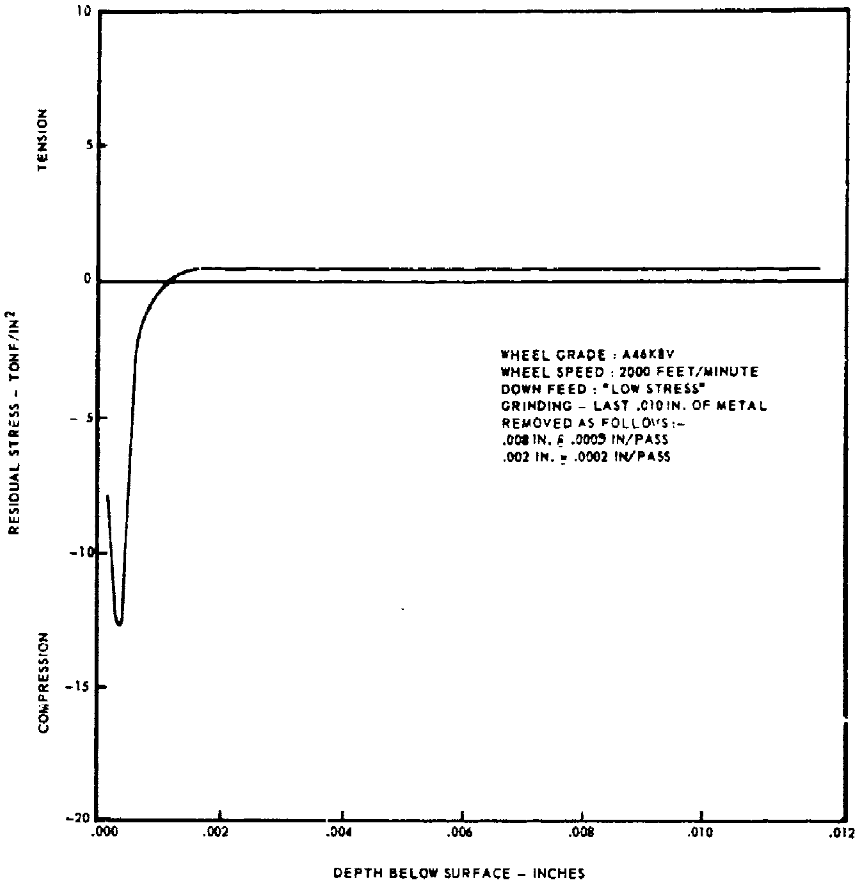


Fig.6 Effect of very gentle grinding conditions on residual stresses of SAE.4340 steel (540 HV). (Field and Kahles[2])

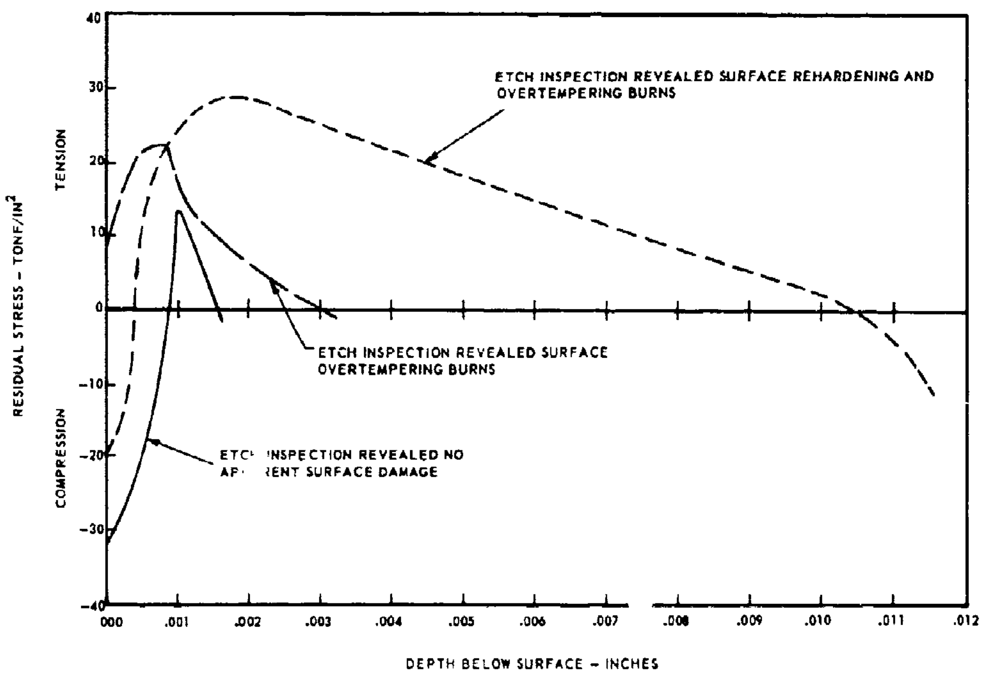


Fig.7 Effect of satisfactory and abusive grinding on residual stresses of BS.S28 steel

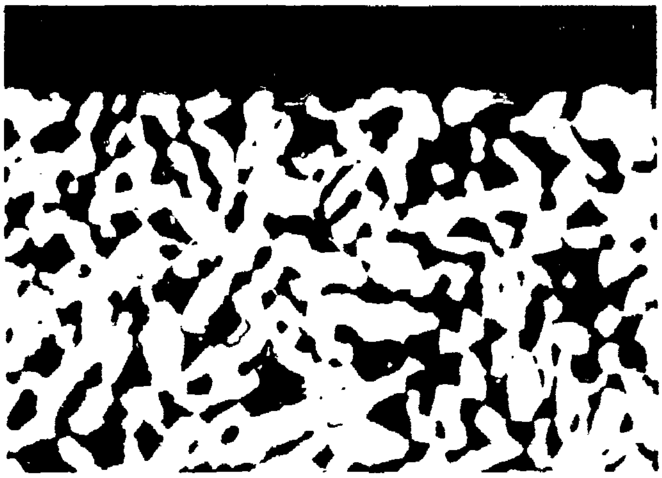


Fig.8 Satisfactorily ground Hylite 51 titanium alloy surface

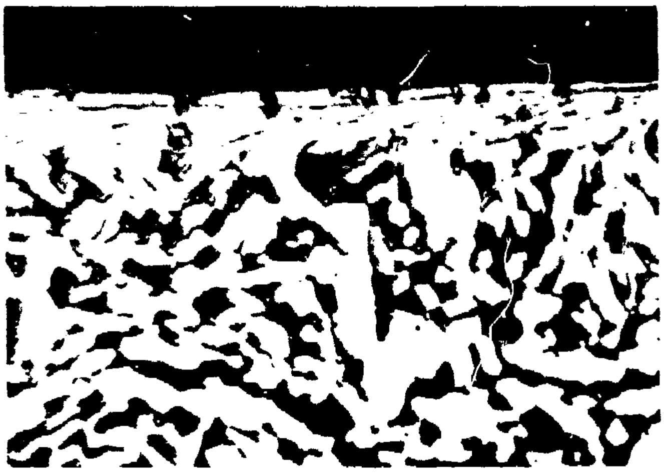


Fig.9 Brittle outer skin at surface of abusively ground Hylite 51 titanium alloy

Fig.10 Re-solidified surface layer, untempered martensite and associated crack at edge of spark-machined hole in NCMV steel

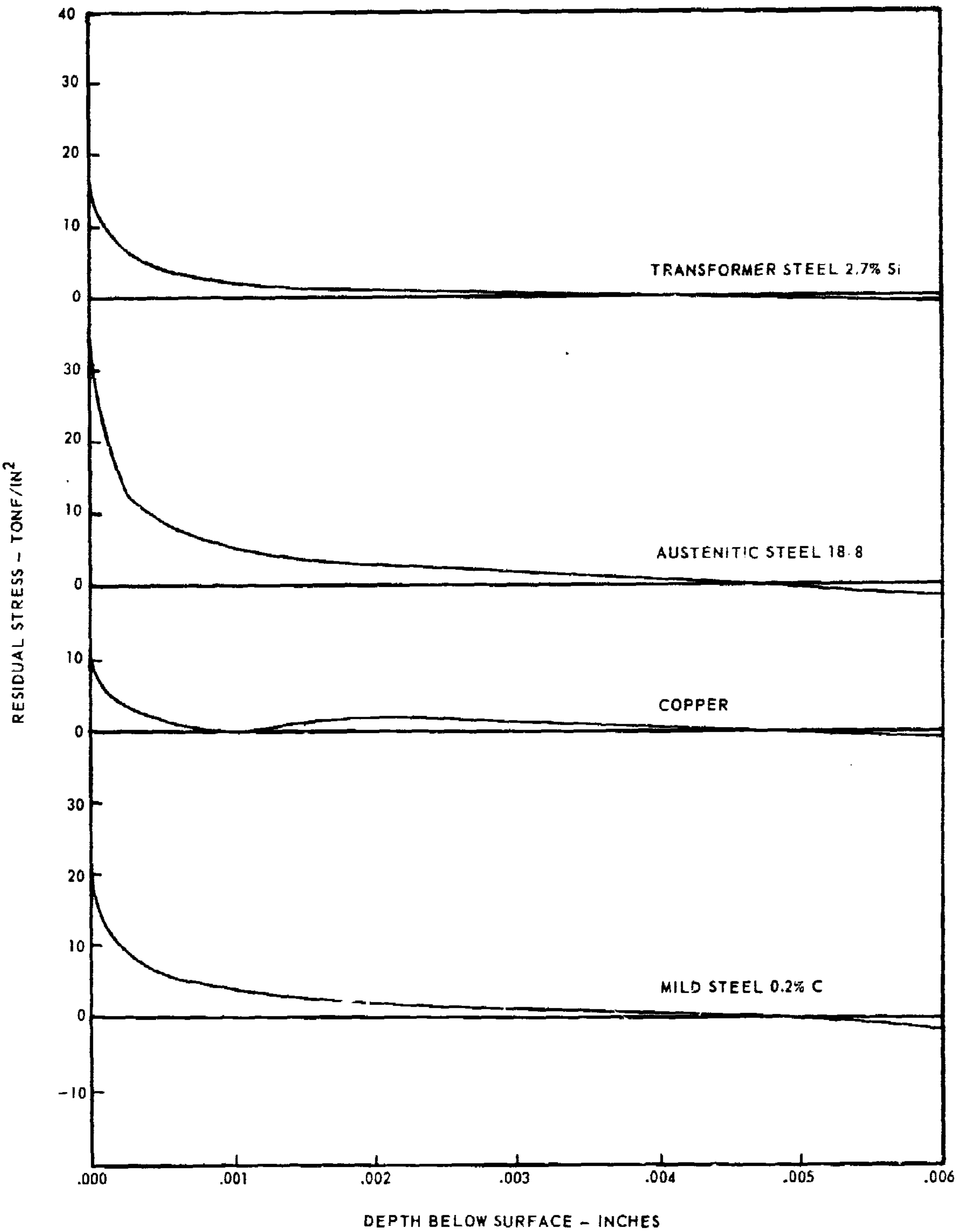


Fig. 11 Residual stresses in spark-machined surfaces (Lloyd and Warren[5])

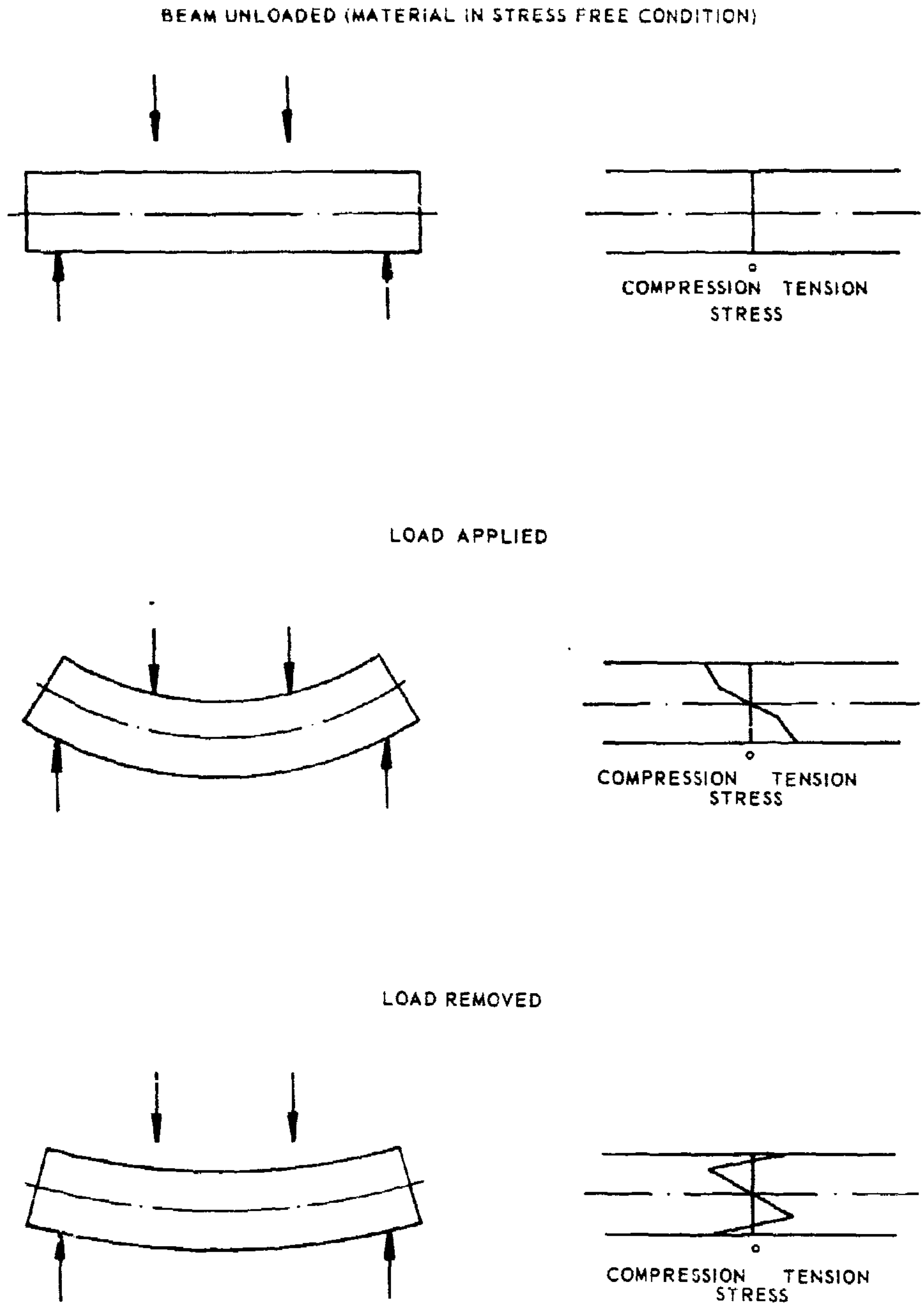


Fig. 12 Stress distribution resulting from the bending of a beam

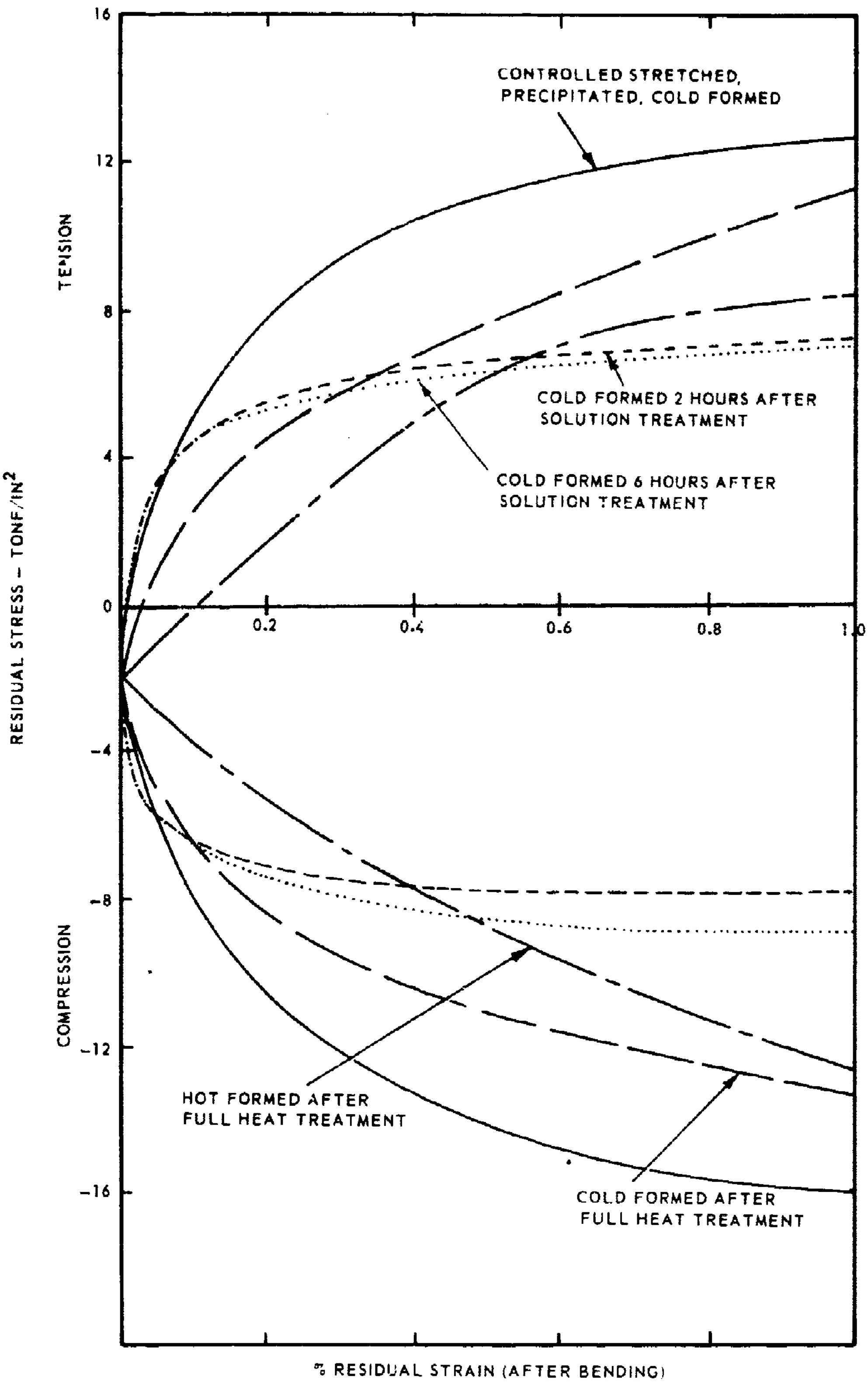


Fig. 13 Relationship between residual stress and residual strain, on bending extruded bars of DTD. 5044 aluminium alloy (Hawkes[7])

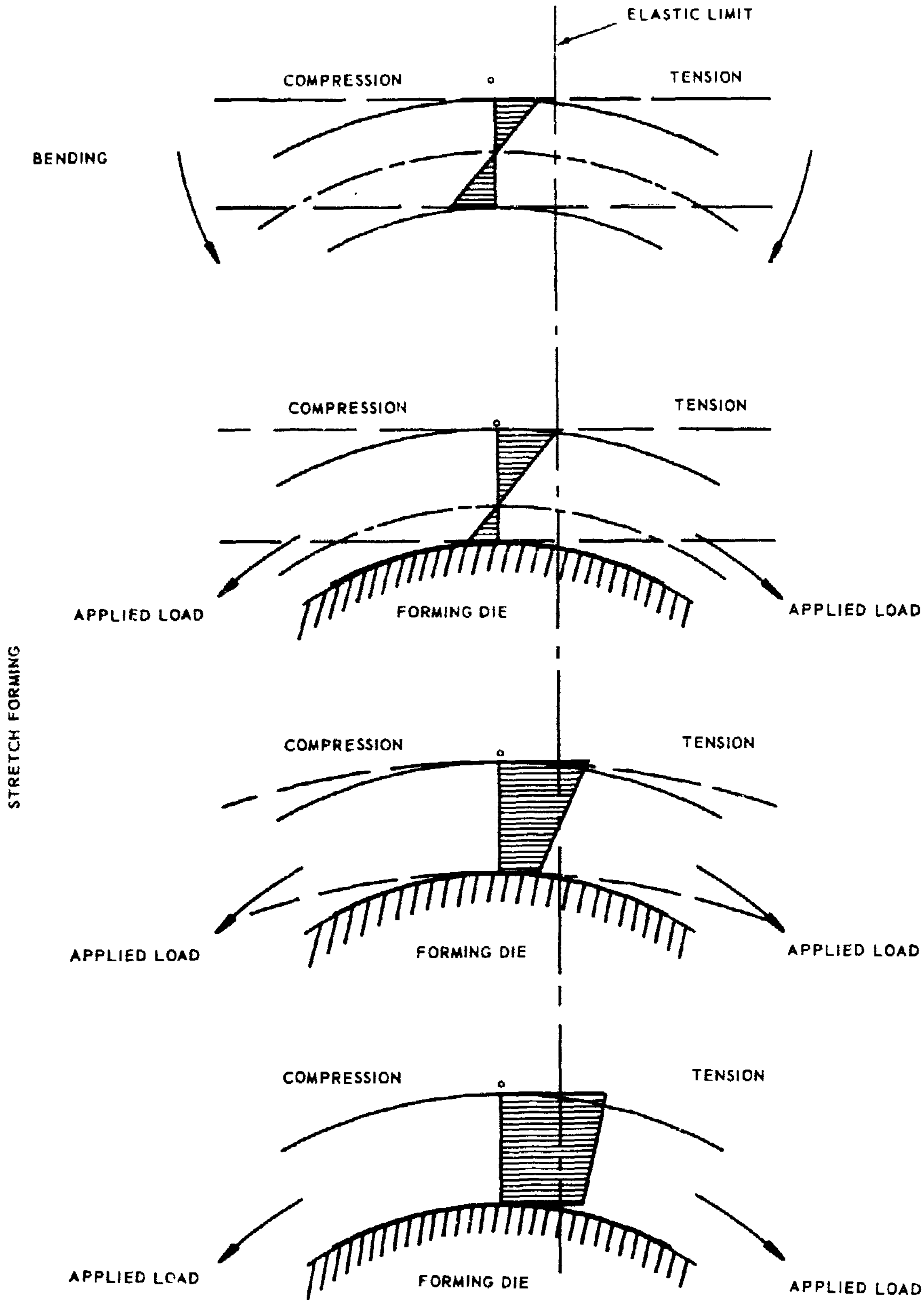


Fig. 14 Stress distribution during stretch-forming (Edwards[9])

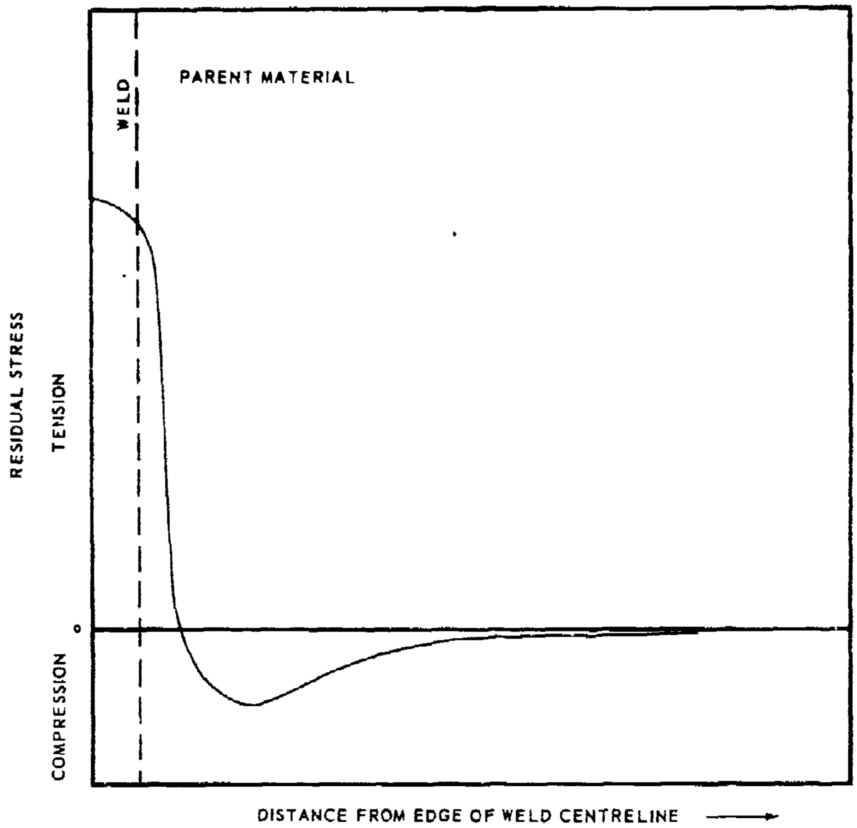


Fig. 15 Surface residual stress distribution in a butt-welded joint

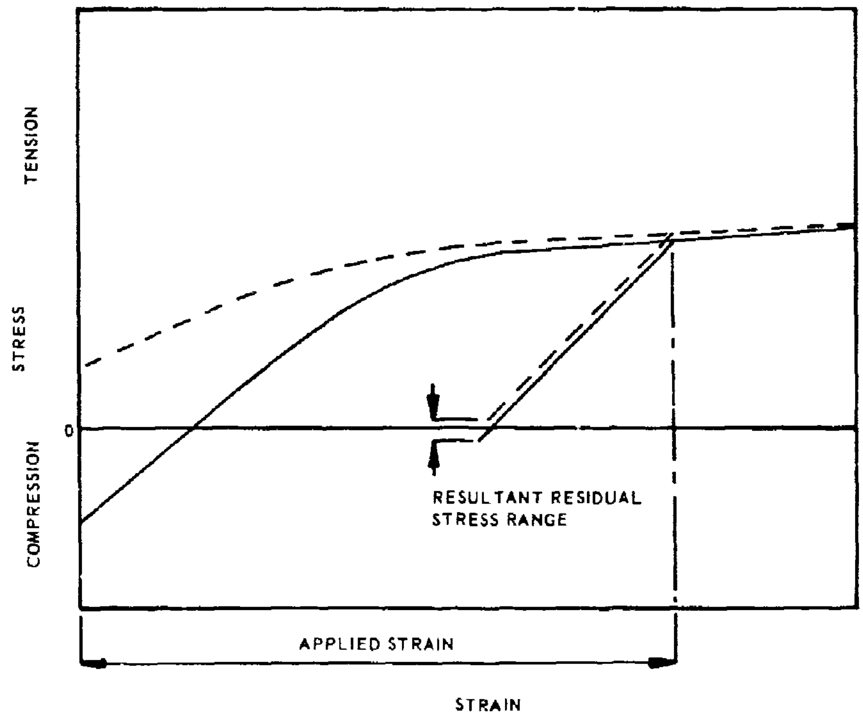


Fig. 16 Residual stress relief by controlled stretching

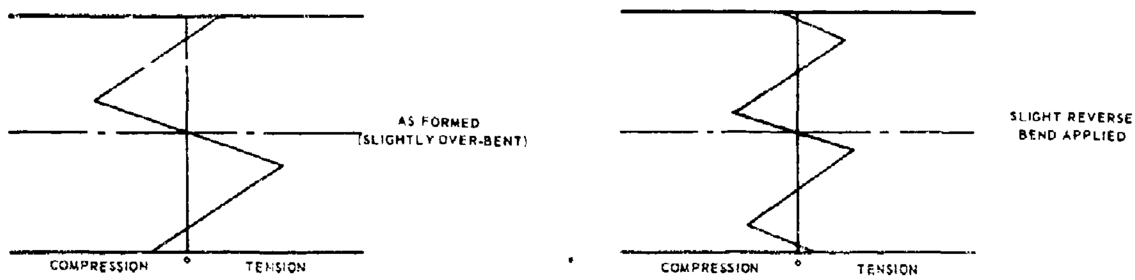


Fig. 17 Residual stress relief by reverse bending

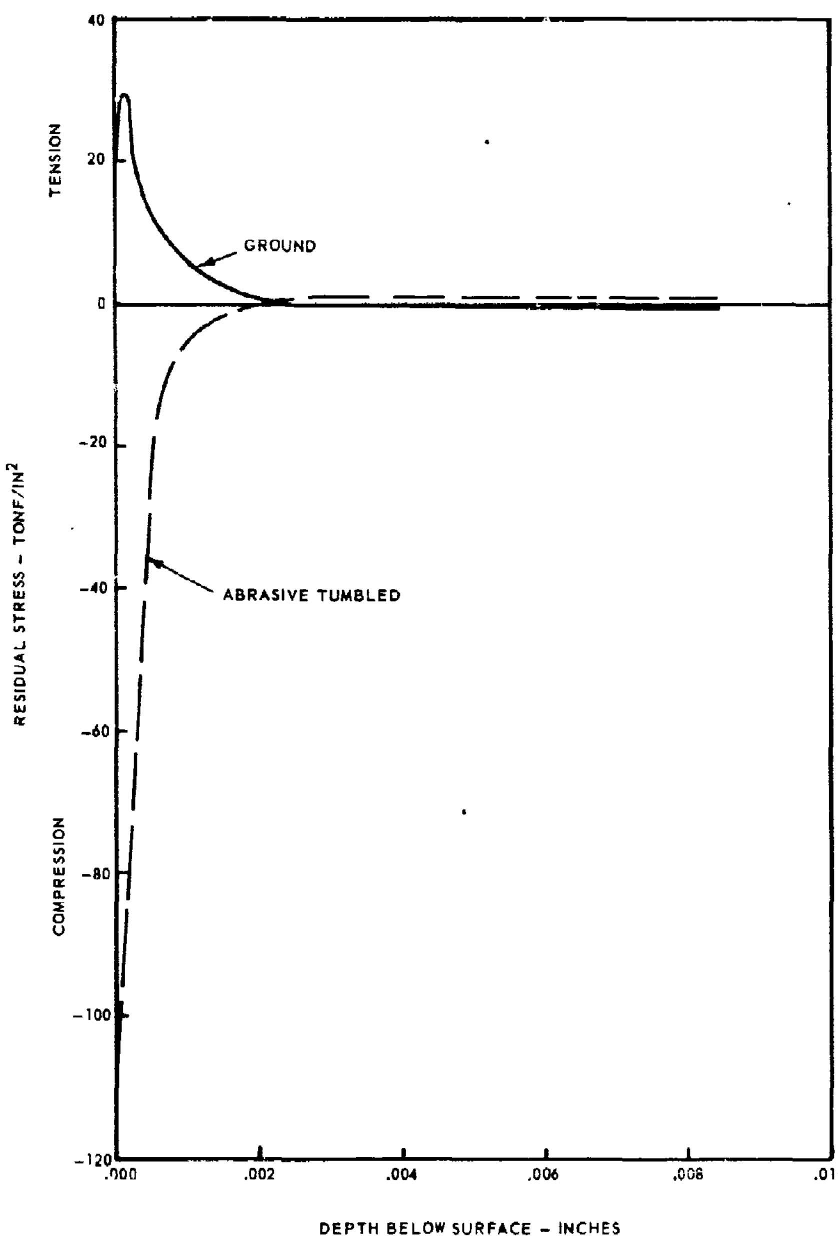


Fig. 18 Effect of abrasive tumbling on residual grinding stresses of AISI. 52100 steel (670 HV). (Tarasov, Hyler and Letner[13])

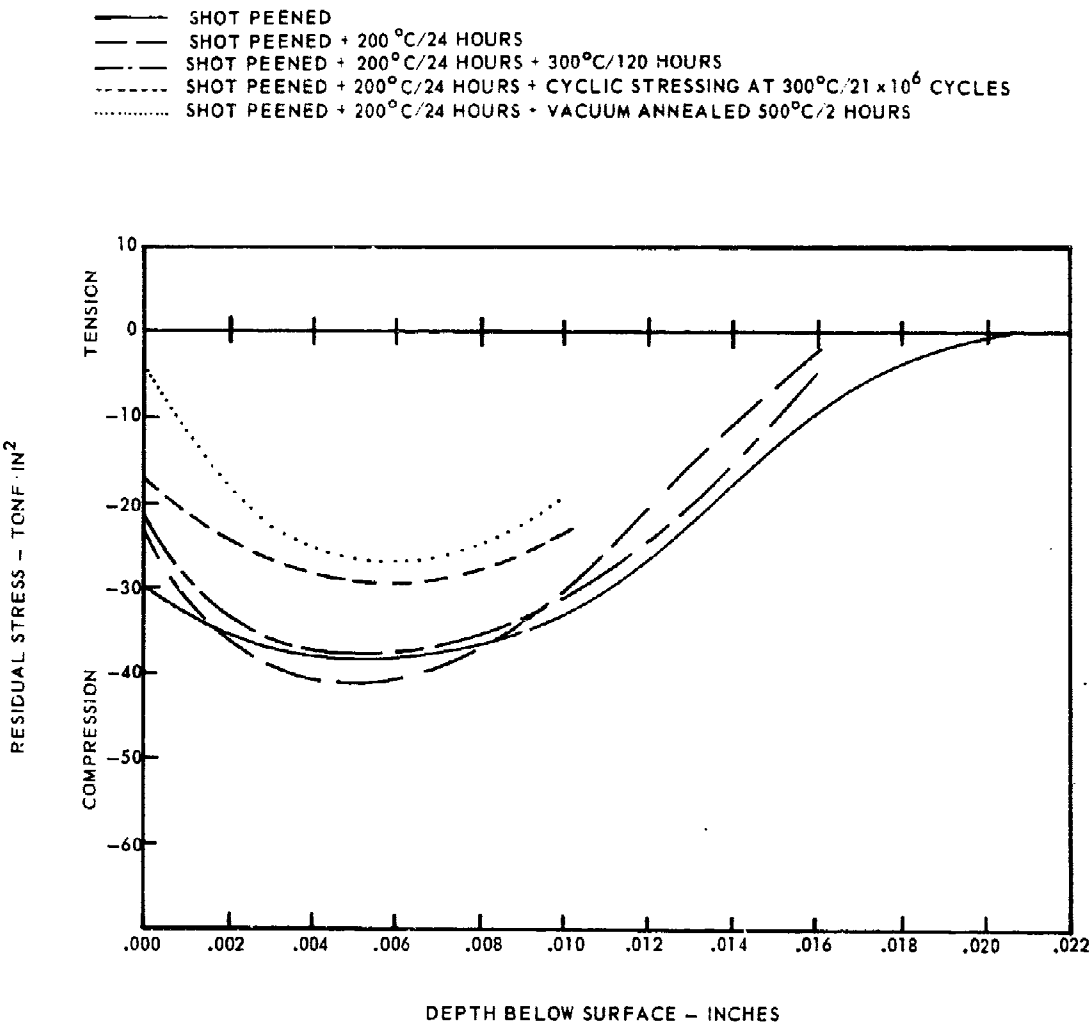


Fig. 19 Effect of cyclic stressing and thermal treatments on the residual stress distribution of BS.S99 steel fatigue specimens. (Westland Helicopters Limited[15])

STRESS CORROSION TESTS ON WELDED CRUCIFORMS OF 7039 - TYPE ALUMINUM ALLOY

by

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STRESS CORROSION TESTS ON WELDED CRUCIFORMS OF 7039 - TYPE ALUMINUM ALLOY

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1. INTRODUCTION

The wrought Al-Zn-Mg-Cu alloys are susceptible to stress corrosion cracking especially where end-grain fibering has been exposed by machining. Thus the exposed ends of plates and extrusions are dangerous areas if stressed above a critical level in the neighborhood of 10 per cent of the yield strength. These facts are well understood in the design and fabrication of such components as airplane landing gears.

When copper is excluded from this alloy system, the material is weldable. This gives rise to the possibility of fabricating structures by weldments composed of plates and extrusions. In such structures, the end grain may be exposed and subject to stress because adjacent welds provide a source of residual stress. It is the purpose of this paper to describe tests which were made to assess the degree of stress corrosion cracking susceptibility under various weld conditions and to point out the relative effectiveness of different protective measures.

2. EXPERIMENTAL PROCEDURE

2.1 Materials

The Al-Zn-Mg alloy tested (type 7039) conformed to the composition shown in Table I. The filler wire composition (type 5183) is also shown in Table I. Mechanical properties of the three tempers tested of the 7039 alloy are listed in Table II.

For comparative purposes, a few specimens of an Al-Mg alloy (type 5083) were also tested. Composition of the alloy is shown in Table I (type 5183 filler was also used).

Alloy 5083 is resistant to stress corrosion cracking unless heated during the final stages of processing, when it becomes extremely susceptible in the short transverse direction. For this reason, tests on 5083 alloy specimens were conducted in the "as received" condition and after heating for seven days at 212°F.

2.2 Fabrication of Weldments

Figure 1 (sketch A) illustrates the arrangement of the plates and the general welding procedure for fabrication of the welded structures. The machine-finished slices removed from the structure are referred to as the individual cruciform specimens. Also shown (sketch B) is the cruciform specimen with the stub leg prepared for test.

The following details supplement the information in Figure 1:

1. All welds are single pass using automatic MIG process, with 5183 alloy filler wire.
2. Interpass temperature - start first weld at room temperature; for remaining welds maintain between 125 to 150°F for each cruciform.
3. Temperature should be taken 3 in. from weld root at plate center on through-plate surface.
4. Amperage - 390; voltage - 28-1/2; speed of travel - 11 in. per minute.
5. Welding machine - inert gas-shielded type (argon); electrode diameter - 3/32 in.
6. Fillet size - 1/2 in.

Cruciform structures were fabricated from plates of each 7039 temper shown in Table II.

2.3 Test Environments

Tests were conducted in three environments. These were the accelerated alternate-immersion test and outdoor exposures in marine and in industrial environments.

In the alternate-immersion tests, specimens were immersed for 10 minutes of each hour in a 3.5 percent (by weight) solution of NaCl, prepared from reagent grade salt and distilled water. The specimens were supported in the test tank so that the bottom surfaces were also exposed to the solution. In the remaining 50 minutes of each hour, the specimens were exposed to the air and apparently dried between cycles. Fresh solution was prepared weekly.

The marine exposure site is located on the coast of southern New Jersey, approximately 500 feet from mean high tide. Here the cruciforms were supported on racks, with the short transverse stubs facing the ocean.

The industrial site is located on the roof of a Frankford Arsenal building (Philadelphia, Pennsylvania). This site was selected because of the abundance and diversity of industrial activity in the area. At this location, the cruciforms were also supported on racks.

3. RESULTS

3.1 Susceptibility vs Stubs Length

The first group of tests was conducted to determine the stub length most susceptible to stress corrosion. One of the through (12-inch long) pieces of each specimen was cut off to produce a stub of 3/4, 1-1/4, 2-1/4, or 3-1/4 inches. Six specimens of each temper were made with each of the above stub lengths. Two of each were tested in alternate immersion, two in the marine, and two in the industrial environment (Table III).

Figure 2 shows the results of these exposures. It is clear that the 3/4-in. stub length is the most susceptible with no susceptibility detected in the 2-1/4 in. and 3-1/4 in. stub lengths even with exposure times over 1600 days.

A typical stub face crack is illustrated in Figure 3.

In a preliminary series of cruciform tests, failures occurred in the short transverse sides of cruciforms both on the through plate and on the 6-in. legs. These failures sometimes preceded failure on the stub face (Fig.4). To avoid this source of stress relief on the stub face, all test specimens were coated with a liquid neoprene paint to prevent attack on any face other than the stub face.

3.2 Comparative Tests on 5083 Alloy

The comparative tests with 5083 alloy 3/4-in. stub, cruciform specimens yielded the results shown in Figure 5. As expected, the normal alloy did not crack with exposures as long as 1,000 to 2,000 days (Table III).

The same material was highly sensitized however by the grain boundary precipitate resulting from 7 days heating at 212°F. Failure occurred in only 2 days in alternate-immersion.

3.3 Tests on Protective Measures

Various techniques were investigated to eliminate cracking of the stub face. These included shot peening, alodining, painting, special heating, and weld overlaying.

Shot peening is an established technique and was tested in conjunction with alodining and painting cycles including paint backing of 2-1/2 hours at 135°F.

Direct heating of the stub by a gas torch was also tested as a means of stress relief; the temperature was raised to 425°F using temperature sensitive paints to control the maximum temperature rise.

It was felt that heating the weld bead directly might act to relieve stress on the stub face; this temperature was controlled to a maximum of 300-325°F using temperature sensitive paints.

Finally, the use of an overlaid weld over the stub face was studied. The overlay alloy was 4043 (Table I).

Results of the stub and weld bead heating methods are plotted in Figure 6. It can be seen that none of these methods is effective.

However, Table IV shows that both shot peening and overlay welding are effective.

4. DISCUSSION

These data clearly indicate that short transverse (end-grain) of susceptible aluminum alloys will crack by stress corrosion when stressed by adjacent welds. The distance of the stub face from the weld is critical in determining susceptibility or immunity.

The tests on protective measures indicate that effective protection is afforded by shot peening or overlaying welding. It is obvious that overlay welding, where design permits, is to be preferred because inspection is simple and the results are assured.

Shot peening is also effective but inspection is difficult and the effectiveness may be lost where abrasion or wear may erode the peened surface.

5. CONCLUSIONS

A 3/4-in. long stub length on a cruciform sample was more susceptible to stress corrosion than a 1-1/4-in. stub. Stubs of 2-1/4 and 3-1/4 inches were immune for the three-year period of this test.

Shot peening and weld overlaying with 4043 alloy were the only effective means tested for preventing stress corrosion of short transverse stubs which were in tension from opposing welds.

TABLE I

Chemical Composition of Materials

Element percent by wt	Plate Alloys		Weld Rod Alloys	
	7039	5083	5183	4043
Zn	3.5 - 5.0	0.25 max	0.25 max	0.10 max
Mg	2.0 - 3.8	4.0 - 4.9	4.3 - 5.2	0.05 max
Mn	0.10 - 0.70	0.30 - 1.0	0.50 - 1.0	0.05 max
Cr	0.06 - 0.25	0.05 - 0.25	0.05 - 0.25	-
Si	0.30 max	0.40 max	0.40 max	4.5 - 6.0
Fe	0.40 max	0.40 max	0.40 max	0.8 max
Zr	0.20 max	-	-	-
Ti	0.10 max	0.15 max	0.15 max	0.20 max
Cu	0.10 max	0.10 max	0.10 max	0.30 max
Others, each	0.05 max	0.05 max	0.05 max	0.05 max
Others, total	0.15 max	0.15 max	0.15 max	0.15 max
Al	Remainder	Remainder	Remainder	Remainder

TABLE II

Mechanical Properties of Plate

Lot No.	Temper	F_{tu}, ksi	F_{ty}, ksi	e, percent in 2 in.
7039 Alloy* (Solution Treated and Aged)				
361	TX	65.2	55.6	12.5
711	T6E121	65.2	56.0	14.0
251	T61	60.5	49.2	12.5
311	T61	60.1	49.8	13.5
5083 Alloy** (Cold Worked)				
-	-	48.5	43.2	14.5

* Properties of specimens taken from direction transverse to rolling direction.

** Properties of specimens taken from direction parallel to rolling direction.

TABLE III

Stress Corrosion Data on Cruciform Specimens Having Unprotected Stubs of Various Lengths

Lot No.	Stub Length (in.)	Alternate Immersion		Marine Exposure		Industrial Exposure	
		F/N	Failure Times (Days)	F/N	Failure Times (Days)	F/N	Failure Times (Days)
5083-H115							
246391	3/4	0/1	OK-1103	0/1	OK-1938	0/1	OK-1939
246391*	3/4	1/1	2	1/1	14	1/1	36
7039-T61							
255251	3/4	2/2	21,718	2/2	609,791	2/2	209,209
	1-1/4	0/2	OK-843	1/2	1011	2/2	244,257
	2-1/4	0/1	OK-843	0/1	OK-1674	0/1	OK-1675
255251	3-1/4	0/2	OK-843	0/2	OK-1675	0/2	OK-1676

F/N = No. of specimens failed/No. of specimens tested

* Heated 7 days at 212°F.

TABLE IV

Stress Corrosion Data on 7039 Alloy Cruciform Specimens Having 3/4-Inch Stubs Protected by Various Methods

Lot	Treatment	Failure Time, Days					
		F/N	NaCl	F/N	Marine	F/N	Industrial
218361-TX	Shot peened stub	0/1	OK-1023	0/1	OK-1017	0/1	OK-1860
	Shot peened, 2-1/2 hr at 135°F	1/1	326*	0/1	OK-1017	0/1	OK-1860
	Shot peened, alodined & painted, 2-1/2 hr at 135°F	0/1	OK-1023	0/1	OK-1017	0/1	OK-1860
	Stub heated to 425°F	1/1	16	1/1	28	1/1	85
	Stub heated to 400-425°F, alodined, painted, 2-1/2 hr at 135°F	1/1	754	0/1	OK-939	0/1	1046
	Four weld beads heated to 300-325°F	0/1	OK-945	1/1	244	1/1	74
	Weld overlay of 4043 on stub	0/1	973**	0/1	OK-1812	0/1	OK-1817
199711-T6E121	Shot peened stub	0/1	OK-1023	0/1	OK-1017	0/1	OK-1023
	Shot peened, 2-1/2 hr at 135°F	1/1	326*	0/1	OK-1917	1/1	349*
	Shot peened, alodined & painted, 2-1/2 hr at 135°F	0/1	OK-1023	0/1	OK-1854	0/1	OK-1860
	Stub heated to 425°F	1/1	91	0/1	OK-953	1/1	170
	Stub heated to 400-425°F, alodined, painted, 2-1/2 hr at 135°F	0/1	OK-975	0/1	OK-1776	0/1	OK-975
	Four weld beads heated to 300-325°F	0/1	OK-975	1/1	76	1/1	17
	Weld overlay of 4043 on stub	0/1	OK-981	0/1	OK-1812	0/1	831**
255251-T61	Shot peened stub	0/1	OK-863	0/1	OK-1665	0/1	OK-1670
	Shot peened, 2-1/2 hr at 135°F	0/1	OK-863	0/1	OK-1665	0/1	OK-1670
190311-T61	Shot peened, alodined & painted, 2-1/2 hr at 135°F	0/1	OK-863	0/1	OK-1665	0/1	OK-1670
255251-T61	Stub heated to 425°F	0/1	OK-857	0/1	OK-1644	1/1	639
	Stub heated to 400-425°F, alodined, painted, 2-1/2 hr at 135°F	0/1	OK-857	0/1	OK-849	0/1	OK-1652
	Four weld beads heated to 300-325°F	0/1	OK-857	0/1	OK-849	0/1	OK-1652
	Weld overlay of 4043 on stub	0/1	OK-843	0/1	OK-842	0/1	OK-1638

F/N - No. of specimens failed/No. of specimens tested
* Failed outside peened area on side of through leg
** Cracked on end opposite weld area.

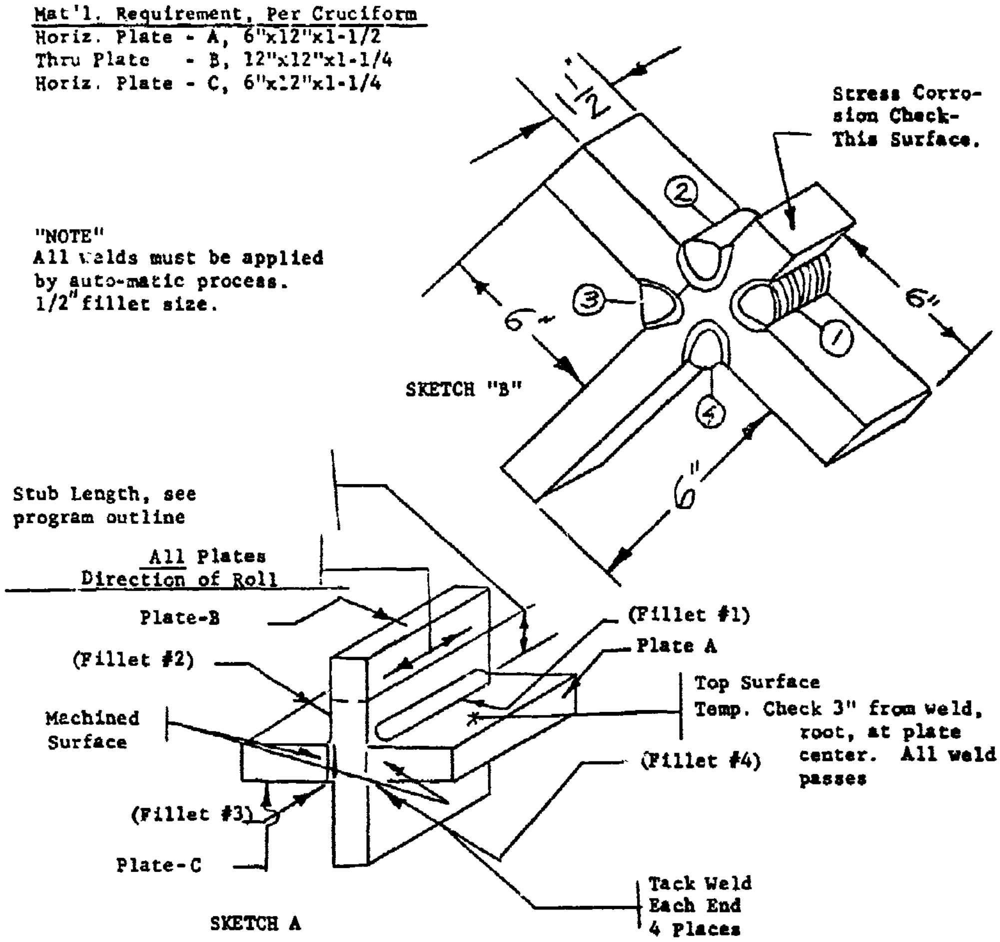

Fig. 1 Design and fabrication of cruciform structure and cruciform specimens.

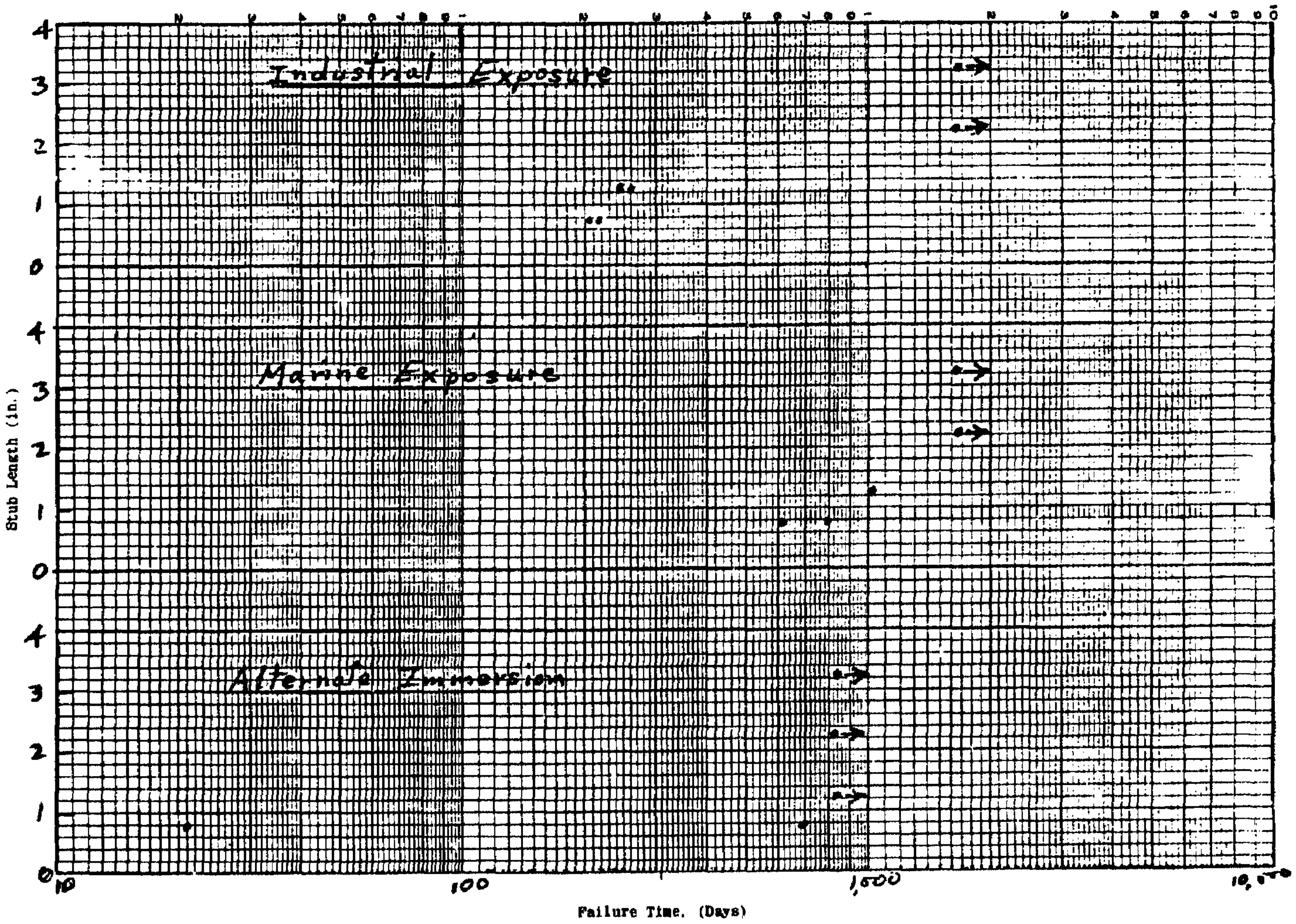

Fig. 2 Exposure results on 7039 alloy cruciform specimens with various stub lengths

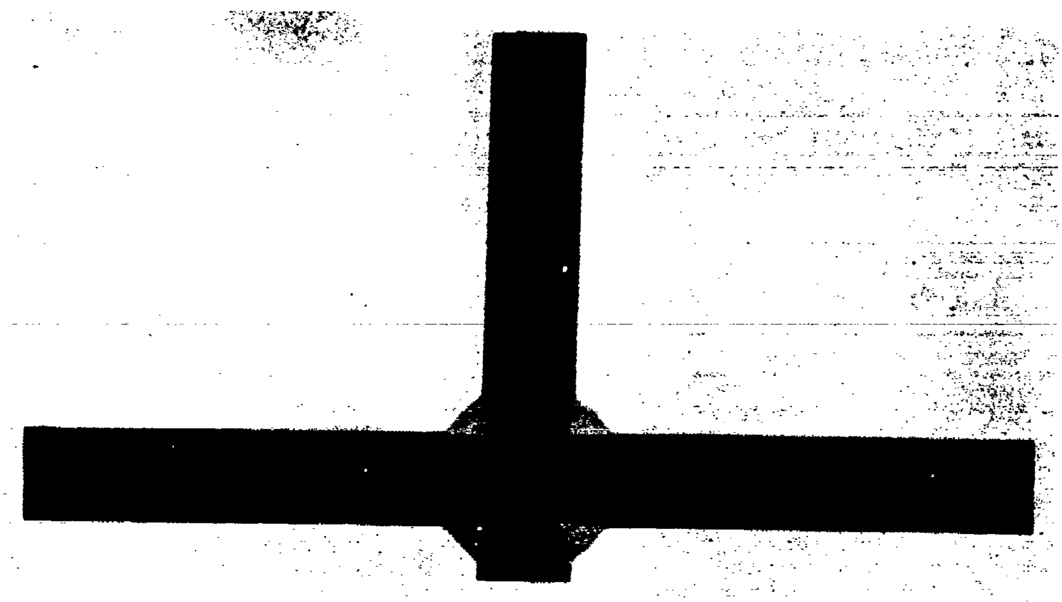

Fig.3 Macroetched 7039 alloy cruciform with through leg cut to 3/4-in. stub length. Original failure occurred after 31 days exposure to an industrial environment. Crack had increased in length at the time of photographing

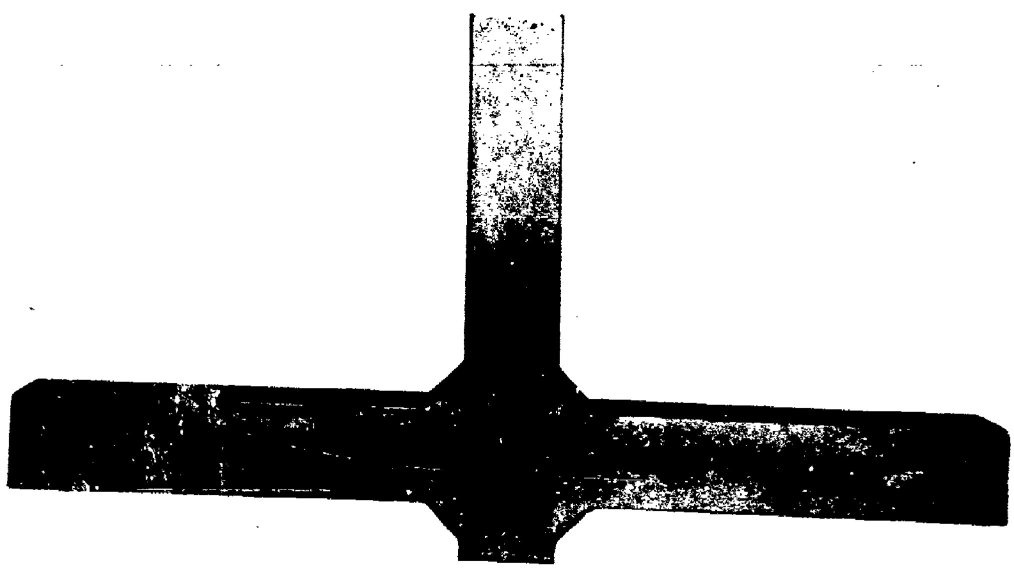

Fig.4 Cruciform of 7039 alloy showing crack which occurred on short transverse side of through plate. Neoprene-coated specimens were used to prevent failure on these surfaces

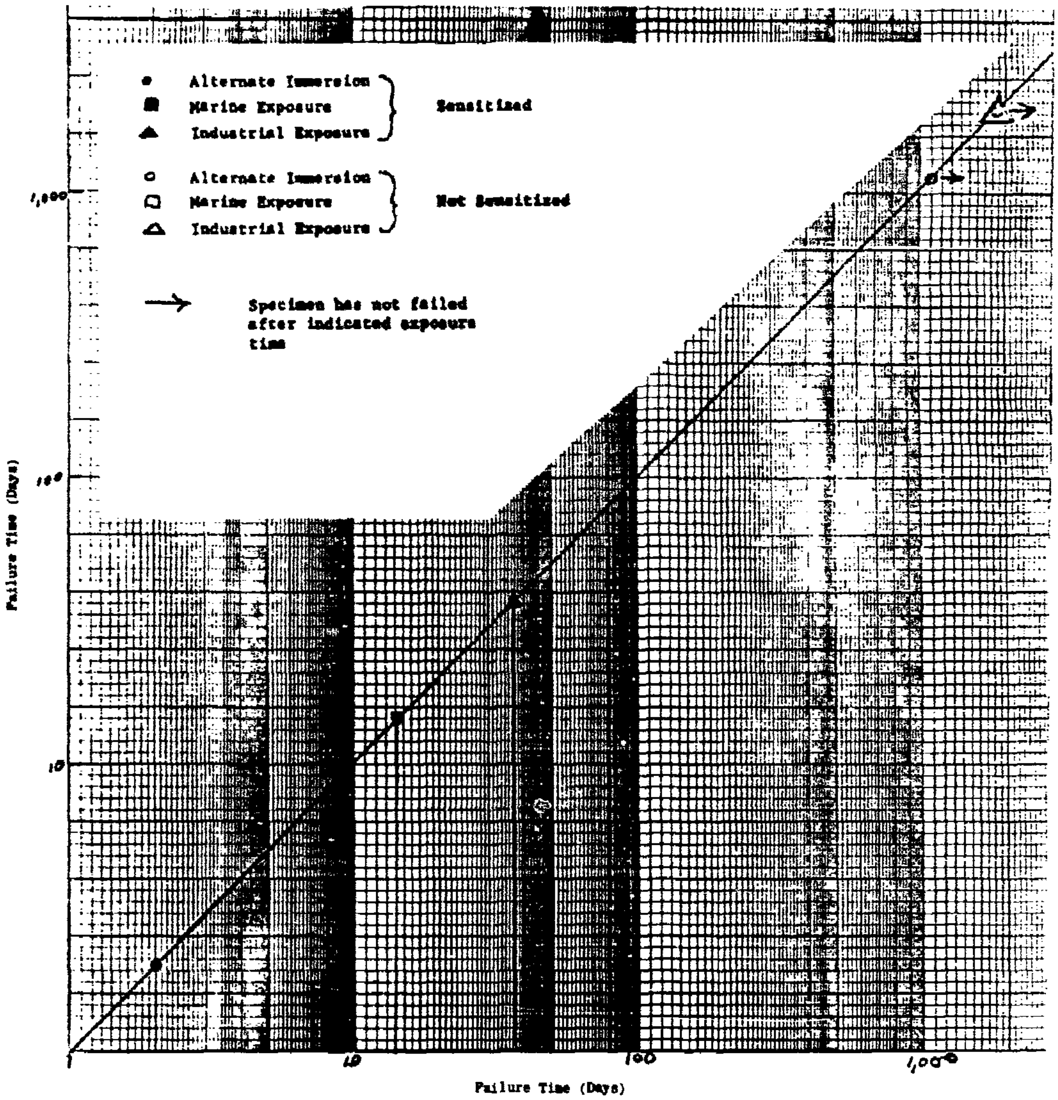

Fig.5 Exposure results on 5083 alloy cruciform specimens with 3/4-in. stub length

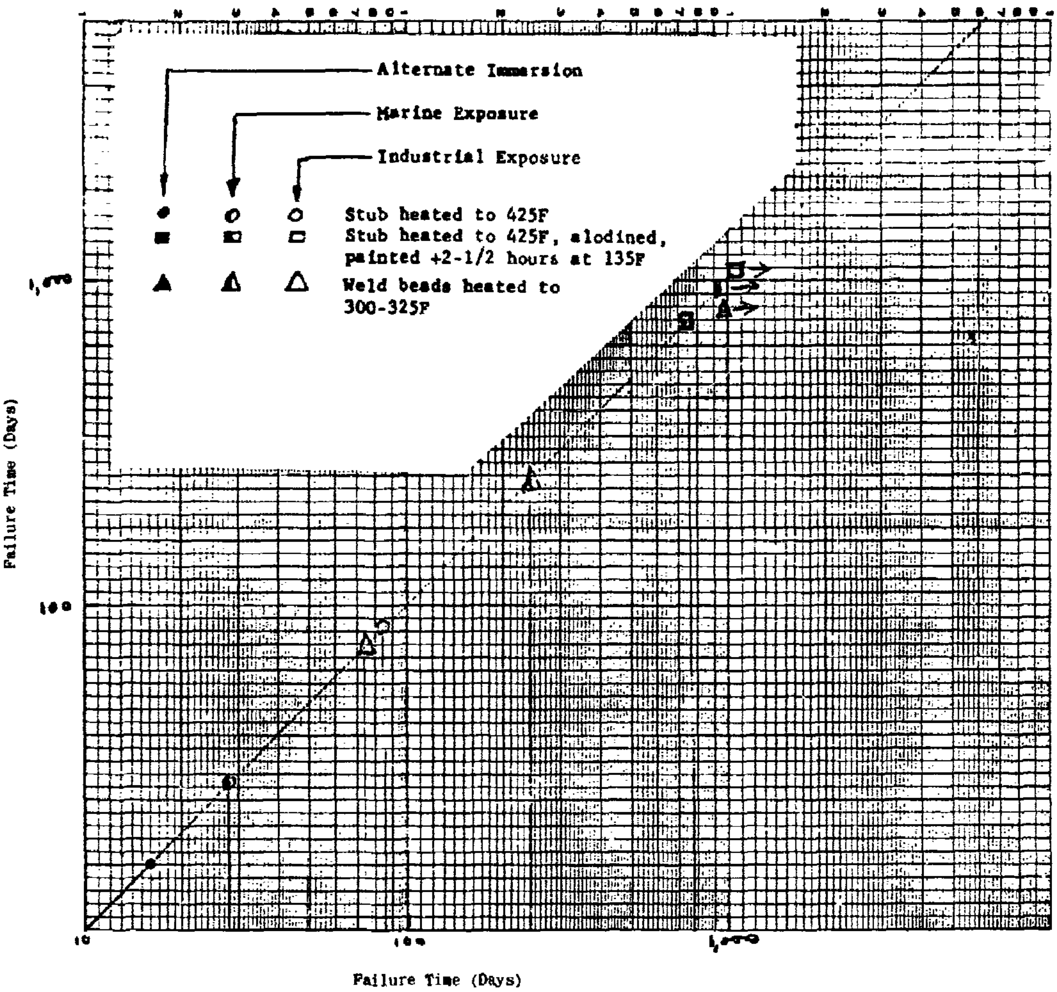

Fig. 6 Exposure results on 7039 alloy cruciform specimens (3/4-in. stub) treated for stress relief by direct heating of the stub face or the weld bead

THE APPLICATION OF N.D.T. TECHNIQUES TO THE PROBLEM OF STRESS CORROSION CRACKING

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Summary

This paper examines the areas where improved nondestructive testing methods could assist in determining the susceptibility of components to stress corrosion, supporting laboratory investigations into the mechanism of stress corrosion and monitoring the progression of cracking during the service life of components to replace the present practice of extrapolating from laboratory simulation tests.

Emphasis is directed towards methods of residual stress measurement which seem closest to a practical test to replace the destructive stress relaxation methods currently used.

1. Introduction

Stress corrosion cracking is one of the mechanisms that can cause premature failure during service of certain engineering components and, as such, it has been a subject of considerable practical concern and laboratory study since it was first recognised in 1886.

Reduced to its simplest fundamentals, stress corrosion, or season cracking, is the initiation and growth of cracks from the surface of a susceptible metal or alloy: The surface must have a high built-in static tensile stress and be in contact with a specific corrodent at the right concentration. The significant metallurgical feature is that the level of stress would not, in itself, be sufficient to cause premature failure and the environment would not, in itself, produce detrimental corrosion in the absence of the stress. A number of reviews of the subject have been published (1)(2)(3).

Tests to evaluate specific materials to check their proneness to stress corrosion cracking are generally carried out on specimens in the laboratory, under applied stress and in a specified environment which either simulates service conditions or is designed to produce accelerated attack (4)(5)(6).

The purpose of this paper is to review how some of the more recently developed nondestructive methods of inspection might be used to predict susceptibility to stress corrosion of a particular component after assembly and prior to being put into service, or to monitor a component for the onset of stress corrosion micro-cracking during service. It is hoped that the paper might stimulate further development work on some of the techniques in order to turn them into reliable procedures that can be used to replace the rather inadequate and, in many ways, unsatisfactory methods based on the laboratory simulation tests.

2. Areas of N.D.T. Application in Stress Corrosion

Fig.1 summarises the full cycle of stress corrosion from component assembly to component failure. It pinpoints some of the features that influence or control the mechanism and indicates the type of test that might be developed to provide a nondestructive examination before or during service, or provide a more positive control during laboratory investigations.

Apart from monitoring the initiation and progression of the cracks leading to ultimate failure the main area of interest is probably in developing reliable nondestructive methods of detecting and measuring the stress level on a metal surface and this is where recent technological progress in N.D.T. looks likely to suggest some suitable techniques.

3. Stress Measurement

Stress in a metal is a non-equilibrium condition caused by some mechanical or thermal deformation of the atomic lattice and a subsequent restraint on its return to equilibrium. There are a number of features associated with stress that need to be appreciated before its effects can be predicted and before its presence can be reliably monitored by any nondestructive test. These are listed in Fig.2.

In stress corrosion, residual stresses as distinct from applied or design stresses, are those most difficult to account for, since they can be caused by a large number of fabrication or metal-treatment processes at various stages during the manufacture of a component (Fig.3). Heat treatment cycles can usually be specified to remove or reduce the level of residual stresses, but they need to be very carefully controlled or a fresh distribution of residual stresses may be introduced as a direct result of the heat treatment.

Destructive methods for determining stress levels have been developed based on careful observation and measurement of the distortion or displacement of cut surfaces or slit tubes and estimating the stresses necessary to account for the relaxation deformation (8)(22). These tests however, can by their nature, only be applied to test specimens of simple geometry and, at best, are only a qualitative guide to the situation likely to exist in a particular component.

3.1 X-ray Stress Measurement

X-ray crystallographic methods of measuring residual stress have been well developed and in many circumstances used to give detailed quantitative data (9)(10). Elastic macrostresses can be detected as a local strain in the atomic lattice which produces a variation in the lattice parameters. By using high-angle back reflection techniques with either film recording or a diffractometer, very small changes in lattice parameter can be measured with high precision. The atomic strains can be related to corresponding stress levels by using appropriate elastic constants to produce quantitative data on the uniaxial or biaxial stress systems characteristic of the surface.

Under optimum conditions with sharp and well resolved diffraction rings, local lattice strains can be measured corresponding to stresses as low as about 3000 psi. However, optimum conditions are the exception rather than the rule in practice and the various factors that limit the accuracy on application of the technique are shown in Fig.4. Microstresses produce diffuse diffraction rings due to local lattice distortion and the degree of broadening is a qualitative measure of this form of induced stress. Measurement is difficult and the technique is only reliable under way carefully controlled conditions.

The author has examined Nimonic surfaces machined in a variety of ways and the degree of X-ray line broadening that was measured is given in Fig.5 (11). The results are not easy to interpret but they suggest in qualitative terms, that the more the "rubbing" action, as distinct from "cutting" action, in the machining operation, the greater the lattice distortion. Fig.6 summarises the different X-ray indications that can be examined in relation to the various manifestations of stress that can occur in practice.

3.2 Ultrasonic stress measurement

Alternative methods of measuring and analysing stress are based on the propagation of elastic waves at ultrasonic frequencies into the material (12)(13)(14)(22). The obvious potential advantage is that the penetration of the ultrasonic energy into a material need not be restricted to the immediate surface as is the case with X-ray diffraction so that bulk stresses within a component can be monitored. Ultrasonic stress determination techniques can be applied in a number of ways (Fig.7).

3.2.1 Ultrasonic birefringence

When an ultrasonic shear wave is generated in a metal in which there are elastic anisotropy characteristics, it will split into two components polarised in the two elastically significant directions. The velocity of these two waves will be slightly different since the elastic constants which characterise the velocity are different. The introduction of stress is one means of developing an anisotropic condition.

The stress dependence of the elastic constants arises as a result of second order non-linearity in the Hooke's law relationship between stress and strain in the elastic deformation range of metals. This results in the so-called elastic "constants" (Young's modulus and shear modulus) varying slightly with stress. The resultant fractional shear velocity difference is given by:-

$$\frac{\Delta V}{V} = \frac{T}{8\mu^2} (4\mu + n)$$

where μ is one of the two linear theory Lamé constants and n is one of the three non-linear theory Murnaghan constants. T is the uniaxial tension in the material (15).

In the case of a stressed metal the significant directions are the principal stress axes so that the difference in velocity of shear waves polarised in these two directions provides a direct measure of the level of stress.

In steel the shear wave polarised parallel to the stress increases as the tensile stress is raised and the shear wave polarised perpendicular to the shear wave decreases so that the difference velocity increases almost linearly with tensile stress. In aluminium the effects of stress on the two shear waves is reversed (16).

The technique is analogous to the splitting of light beams in photoelastic materials where the permittivity is modified by stress and produces the well known birefringence effect which has been used to measure applied stress distributions in transparent models (17) or in surface coatings(18). In view of this similarity, the ultrasonic effect is sometimes referred to as ultrasonic birefringence, acoustoelasticity or sonaelasticity. An important distinction is that there is a difference of 3 orders of magnitude between the wavelength of ultrasonic waves and light waves so that a stress generating, for example, a thousand light fringes in a photoelastic material will only cause one cycle of phase shift between two ultrasonic shear waves (16). In practical situations the change of velocity caused by a relatively high applied stress is usually only a few parts in 10^4 so that most of the experimental development work has concentrated on methods of measuring these small changes reproducibly and with sufficient accuracy (15).

Unfortunately other structural features apart from stress can produce a similar ultrasonic birefringent effect. In particular, preferred grain orientation in a material which is basically aelotropic can introduce a velocity difference far in excess of that due to any applied stress(19). A suggestion has been made that in steel, where this effect can be pronounced, a distinction between the causes can be made by making measurements at different frequencies (20). In aluminium, the effect of grain orientation would not be expected to have such a masking effect since the basic crystal is more isotropic.

The ultrasonic birefringent technique has shown that completely nondestructive methods of measuring stress are now possible, but some of the cautionary statements of those who have had direct experience of its application should not be overlooked:-

Smith (15) 1963

'Extended research on ultrasonic shear-wave birefringence in polycrystalline solids should lead to interesting information about internal stresses and preferential grain alignment, but it is apparent that considerable difficulties will be met in distinguishing between the two effects.'

Masabuchi (22) 1965

'At the present stage, the ultrasonic technique is a laboratory technique that works only on simple stress fields. Efforts must be made to develop the technique so that it can be used on metal

structures containing complex residual stresses.'

Kammer (21) 1967

'Recent experimental results with ultrasonic waves offer no grounds for optimism that their practical application to this problem will mature in the immediate future ..... . They (the results) make abundantly evident the sobering complexity with which it is necessary to cope before the effects of residual stress on ultrasonic waves can be meaningfully interpreted.'

Benson et al. (12) 1968

'Although the methods have been developed to a useful state for application to practical problems of stress analysis, further study is warranted to improve the spatial resolution of the measurements as well as to further define the effects of grain orientation.'

Crecraft (16) 1968

'The problem of detecting residual stresses in the general case is still not solved, but it seems that we may have a solution for some cases. This remains to be proved.'

Since stress corrosion is initiated at a surface, the advantages of bulk stress measurement are perhaps not as great as they are in other metallurgical situations. In fact since the ultrasonic birefringent method measures the net stress across the section the results will often be misleading since high surface stresses will, in many cases, be completely masked by the compensating stresses of opposite sign in the bulk of the material below the surface through which the ultrasonic beam will also pass.

3.2.2 Ultrasonic surface waves

An alternative ultrasonic technique developed specifically for surface stress determination is based on velocity measurement of Rayleigh waves generated to propagate along the surface. The technique has successfully been applied to stressed aluminium surfaces (12). The ultrasonic energy in this mode of propagation is restricted to a layer approximately one wavelength deep ($\sim$ 1 mm for 3 MHz waves). Special probe assemblies have been developed to make the technique less of a laboratory tool and the results indicate that the velocity variation is linear with stress and that the velocity increases with increasing compressive stress.

3.2.3 Ultrasonic goniometry

A further potentially important stress-measuring technique is based on precise measurement of the critical angles for reflection of ultrasound energy at the metal surface when immersed in a liquid.

At the point of incidence there is a partition of energy between the reflected and refracted waves, but at the two critical angles when the compressional and shear refracted waves are 90° the surface acts as a pure reflector. At the third critical angle, when Rayleigh waves are launched into the surface the intensity of the reflected energy at the incident point falls to a very low value as the surface wave dissipates energy into the water along its path. Detecting and measuring these sharp discontinuities in reflected intensity provides a means of monitoring the elastic characteristics, virtually at a point on the surface. Non-linear elastic theory, invoking the higher-order elastic constants, again predicts a stress dependance on the angle at which the critical energy partition effects occur.

Bradfield (24) pioneered this technique, referred to by him as ultrasonic goniometry, and has used it successfully to observe stresses in steel surfaces and in toughened glass (25). Others are actively continuing the applications development of this technique (26).

3.2.4 Ultrasonic beam interaction

Another manifestation of non-linear elastic theory is the interaction of two ultrasonic beams to generate a third beam, whose point of generation and direction of propagation can be varied by suitably orientating the two incident beams (27). This technique, perhaps of limited interest in stress corrosion, has particular relevance to the case where some internal structure variability needs to be monitored without the results being significantly influenced by surface effects.

3.3 Magnetic stress measurement

A special situation arises with ferromagnetic materials since modifications in their magnetic properties can be caused in some circumstances with stress. The majority of the experimental data so far reported is concerned with studies of applied stresses but there seems no reason why the techniques could not be extended to determination of residual stress.

In summary, the techniques suggested are:-

3.3.1 Magnetostriction effect

When a magnetic sample is loaded in the presence of a biasing magnetising field, the magnetic flux density in the sample, measured by a pick-up coil, will vary with the level of applied strain (28).

As with other stress-measuring techniques other variables can influence the result, and in particular composition, degree of cold work and heat treatment state all affect the magnetostrictive properties.

3.3.2 Magnetoabsorption method

In this technique the sample is placed in a slowly-cycling biasing magnetic field and measurements are made of the changes in energy absorbed from the field of a radiofrequency coil placed in close proximity to the metal under stress. Preliminary results have been obtained with nickel wire specimens under conditions of applied stress (29). The technique is reported to have the advantage that it can be used to distinguish tensile and compressive stresses: materials with a positive magnetostriction constant show increasing magnetoabsorption with tension and vice versa with compression, whereas with a negative constant the results are reversed. Furthermore the effects depend on the orientation of the biasing magnetic field to the direction of principle stress.

3.3.3 Barkhausen effect

Since the hysteresis loop is a result of micro-movements of domain boundaries it can be considered as a discontinuous curve. By using a small pick-up coil very small voltage signals can be detected as each discontinuity occurs during the hysteresis cycling. When amplified this Barkhausen 'noise' signal, first observed by Barkhausen in 1919, can be used to determine structural characteristics of the material (30). A recent study of the effect (31) suggests that the method can be used to measure residual or applied stresses and that the magnitude is a guide to the sign of the stress. The suggestion was also made that a magnetic coating applied to materials might enable stress measurements to be made in this way even though the material under test is not itself magnetic.

4. Monitoring stress corrosion nondestructively

Apart from the problem of measuring residual stress to assess the susceptibility to stress corrosion cracking, there are a number of other aspects of stress corrosion which might be monitored more successfully during the service life of a component (see Fig.1).

Acoustic emission is an extremely sensitive means of detecting crack initiation and propagation(32). The rupture of the lattice at the crack tip is a significant source of stress wave emission and the higher the internal tensile stress the greater the intensity of the emission (33). A wider application of this technique should make much earlier detection of stress corrosion damage possible.

Both the attenuation and reflection of ultrasonic waves provide a very sensitive method of detecting the initiation of surface cracks, and they have the advantage that the ultrasound can be launched into the surface some distance from the corrosion area (34).

The currently available micron-sized point source X-ray sources open up the possibility of 'dynamic' techniques of radiography (35). Since the source and recording film can be separated by a large distance (30-50 cms) with virtually no loss of definition, monitoring of corroding surfaces should be possible - at least on a laboratory scale - to study the onset of cracking and the mechanism of crack propagation through the thickness of the specimen in a very detailed way not previously possible with conventional radiographic techniques.

5. Conclusions

A number of very promising nondestructive testing techniques are currently being developed or assessed that have immediate potential application in the field of stress corrosion testing. None is a panacea and a considerable amount of careful laboratory investigation and calibration is required before these basically more reliable and more positive techniques are used with confidence to replace the existing empirical simulation tests.

Acknowledgement

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Figure 1

Before initiation of stress corrosion	During stress corrosion		At component failure
	At initiation	After initiation	
1. <u>High tensile stress in surface</u> X-ray crystallography Ultrasonic stress determination Magnetic stress determination Stress relaxation tests (destructive)	1. <u>Rupture of protective film</u> Acoustic emission 2. <u>Initiation of surface cracking</u> Acoustic emission Ultrasonic surface waves	1. <u>Growth of crack network</u> Acoustic emission Ultrasonic surface waves (attenuation or reflection) High definition radiography Internal friction Electrical resistivity (eddy currents or microwave cavity)	1. <u>Crack penetration</u> Penetrant testing Pressure or leak testing Radiography
2. <u>Modification of structure causing susceptibility</u> X-ray microscopy Optical microscopy Electrical resistivity	3. <u>Electropotential modifications</u> Solution potential measurements		
3. <u>Environment</u> Analytical procedures			

Figure 2

Stress Variables

1	Origin of stress	Applied, external, operational or design stress	Internal, residual or fabrication stress	
2	Type of stress	Tensile stress with atoms being pulled apart, or restrained at too great a separation. Most significant for stress corrosion	Compressive stress with atoms being pushed together or restrained at too small a separation	Shear or torsional stress with lattice being sheared or, twisted
3	Volume of stress	Large volume elastic macro stresses, uniform or smoothly varying over distances large compared with grain size	Granular elastic stresses varying from grain to grain as a result of plastic deformation partly relieving the stress - the degree of stress relief depending on grain orientation if the material is anisotropic	Microstresses at microstructural features in the metal; such as inclusions, crack tips and other local stress raisers. These stresses can vary in magnitude considerably within grains
4	Uniformity of stress	Uniform across a section - as is possible with residual hoop stress in a ring section or with design stresses	Varying across a section to give a mean stress of zero as is the case with machining or surface treatment stresses	
5	Method of application of stress	Under conditions of constant load - more appropriate to operational stresses	Under conditions of constant strain - more appropriate to fabrication stresses	Under intermediate conditions - general case in service which cannot readily be simulated in laboratory stress corrosion tests (7)

Figure 3

1	Casting	Caused by differential cooling of the melt and preferential surface solidification. Generally present as elastic macrostresses
2	Basic metal forming processes (rolling drawing, extrusion etc.)	Stresses introduced by deformation of the metal and modified by extent and direction of plastic flow
3	Machining (milling grinding, turning etc.)	Surface stresses, often tensile, Introduced by local thermal gradients and surface deformation and extending perhaps to a depth of 150 μm
4	Assembly operations (bolting, riveting, press-fits, shrink-fits)	Very localised stressing and extremely difficult to detect
5	Welding	Largely caused by thermal effects and structural transformations. A complex stress pattern can be introduced in the weld and the adjacent heat affected zones
6	Heat treatments	If a quench is included in the heat treatment cycle tensile stresses can be set up in the surface layers due to differential volume changes associated with rapid cooling
7	Surface deformation treatments (shot blasting, sand blasting, peening etc.)	These can produce cold-work deformation and introduce compressive stresses in the surface
8	Surface diffusion treatments (carburizing, nitriding, aluminising etc.)	These distort the surface layers of atoms and tend to introduce residual compressive stress since the interstitial atoms introduced cannot produce equilibrium lattice expansions
9	Microstructure effects (phase changes, martensitic transformations, segregation, diffusion, slip lines, phase precipitation, etc.)	These can all introduce local microstresses

Figure 4

Limitations of X-ray parameter method of stress measurement

1	Only immediate surface stress is measured since the X-ray penetration influencing the diffraction intensity is restricted to approximately 5 μm. This could be an advantage for stress corrosion studies
2	Only fine grained metals can be examined. Coarse grained samples give a 'spotty' diffraction ring which cannot be accurately measured
3	High accuracy requires a high-angle back reflection of the X-rays and a suitable combination of X-ray monochromatic wavelength and lattice spacing may not be available
4	In any particular case, all of the stress measurements are deduced from a single diffraction ring and this implies that all of the information comes from a limited number of grains and only those with a particular and well-defined orientation to the stress. If plastic deformation has occured the value measured could thus be in fact a 'granular' elastic stress
5	Other factors apart from stress-induced strain can influence lattice parameters. In particular, atomic migration to or from the surface as the result of a heat treatment or surface treatment could result in quite erroneous estimations of stress
6	Only metals and alloys which give a resolved diffraction ring can be examined. Many high strength steels, for example, give very diffuse back reflection diffraction rings which cannot be measured accurately
7	X-ray diffraction is not widely used as a 'shop-floor' technique, but there is current interest in improved portability by using isotopic X-ray sources(23)

X-RAY LINE BREADTH MEASUREMENTS

NO.	MACHINING OPERATION	LINE BREADTH - AB - (RADIANS/1000)
1	REAMED SURFACE	103
2	TURNED SURFACE	85
3	PLUNGE GROUND SURFACE	63
4	MILLED SURFACE	62
5	TURNED SURFACE AFTER POLISHING	58

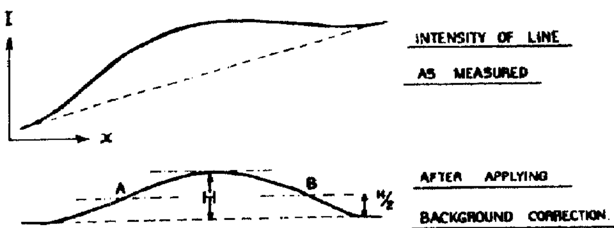


Figure 5

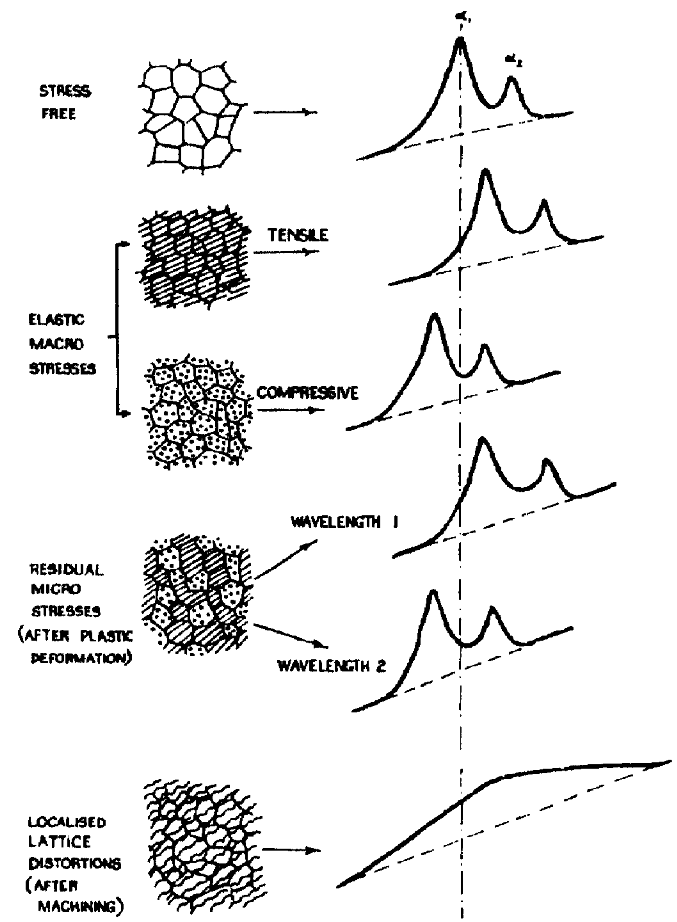
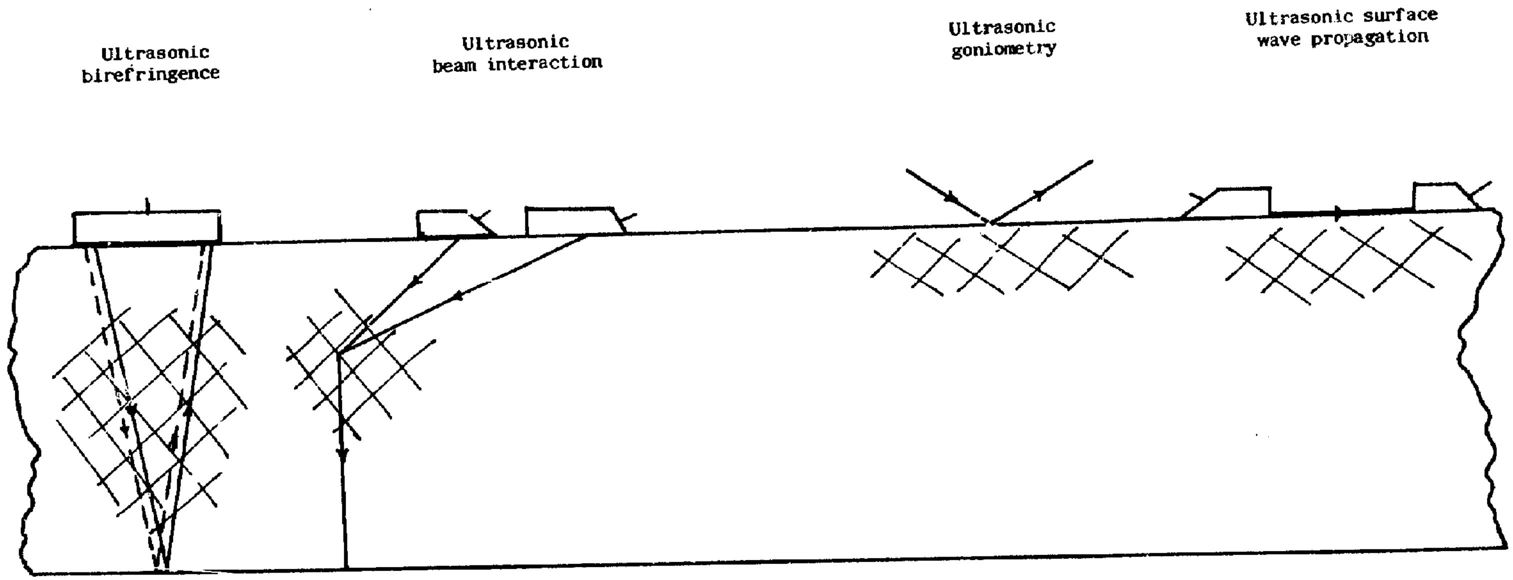


Figure 6

Figure 7


Ultrasonic
birefringence
Ultrasonic
beam interaction
Ultrasonic
goniometry
Ultrasonic surface
wave propagation

THE CHOICE OF MATERIALS

by

G.B. Evans, F.I.M. F.W.Inst. F.R.Ae.S. C.Eng.

SUMMARY

The object of the paper is to suggest one of the means of developing the design of a part, as it influences the choice of material, which can lead to manufacture and service operation without stress corrosion cracking.

The paper draws attention to the different degrees of sensitivity to stress corrosion cracking which can arise in material from small differences of composition heat treatment and coldwork. This emphasises the importance of careful consideration and control at all stages.

An attempt is made to obtain an overall appreciation of the importance of stress corrosion and the other important aspects such as fatigue and it is suggested that all the features necessary to avoid stress corrosion cracks assist with the other problems also, and are good practice even when there is a material of high stress corrosion resistance which would fit the design need.

1. Introduction

The behaviour of the main groups of alloys used in aerospace structures has been outlined in the preceding papers.

Other papers have shown that although a corrosive environment will almost certainly exist, the chances of failure by stress corrosion as has occurred in the past, are reduced to a low order of risk if tensile residual stresses from fabrication and assembly are kept to a low order.

Speakers have discussed the manner by which these stresses can be determined and ways of reducing them.

A survey has shown (1) that although the relatively few tests made with organic protection have not given confidence in their ability to prevent stress corrosion there is nevertheless, a body of opinion arising from operational experience which supports the value of proper paint protection.

Mr. Coles paper (2) in this symposium discusses the influence of protective methods in more detail.

It is with this background and experience that the choice of material for a given purpose can be made.

2. The Origin of Stress Corrosion Failure

For the purpose of discussing the choice of materials it is suggested that the origin of stress corrosion failure can be separated into two categories :-

1. Those failures induced by inherent residual stresses associated with fabrication.
2. Those failures induced by operational stress, i.e. design loads.

The general belief is that manufacturing stresses have been the prime cause of unexpected cracking and it is in these areas therefore, that this paper will concentrate.

It does seem likely however, that with more efficient use of materials in structures that the second cause may become of increased importance. Clearly, if there is a threshold stress beyond which failure will occur by stress corrosion, then this must be considered in the design criteria.

Unfortunately we do not have comprehensive design data to work from. The importance of Dr. Piper's survey (3) is thus emphasised.

In most cases, provided the direction of the applied stress relative to the grain flow is taken into account, the other design criteria and especially the fatigue requirements are likely to ensure that the operational stresses will not promote stress corrosion failures.

This may not be true for pressure vessels and for which therefore, fatigue may not be the controlling feature of design.

3. The Problem in Perspective

Although this symposium is directed at the prevention of stress corrosion cracks, it is necessary when considering the choice of material to view the problem in perspective.

Obviously the failure which gives one the most worry is the one which has just occurred and which is occupying ones full attention, no matter what its manner of failure or origin.

Taking a broader view, the largest number of structural failures on civil aircraft are initiated by fatigue mechanism, perhaps up to 80% of all reported accidents. Stress corrosion only features as one of the several other causes in the remaining 20%.

The proportions of stress cracking troubles in military aircraft might be expected to be higher as in general, a large proportion of military machines are of relatively short life (flying time, not elapsed time), and consequently people appear to consider fatigue is of less significance.

The author's experience would suggest that this is not the case, and that the relative frequency of the different methods of failure are little different from those quoted for civil aircraft.

It is evident from the experience with large rockets (4) that the design criteria are different and probably fall into the second of my two categories of stress corrosion causes. Namely, very high permanent static stress applied for long periods as a feature of the design.

Thus, when considering the choice of material several overall points would seem to provide a background from which to commence the analysis for selection.

1. If due regard is given to the control of unfavourable residual stress, then the stress corrosion behaviour of the material chosen is not as significant in most applications as might appear to be the case.

2. Unfavourable residual stresses can reduce the predicted fatigue performance.

3. All other requirements of design and life expectancy must be taken into account, they are not subsidiary to the stress corrosion performance.

4. When the control of unfavourable residual stress can be coupled with good protective treatments, stress corrosion will cease to be a worry.

5. The fatigue, ductility and crack propagation behaviour of most materials are poorest in the short transverse direction. This aspect must be watched in all design. Stress corrosion, which is also usually worse in this direction, is not the sole reason for concern.

6. It is therefore very good practice to take all these considerations into account even if the material is believed to be inherently resistant to stress corrosion.

7. The majority of the structure in the past has been of high strength aluminium alloys, consequently stress corrosion is more usually associated in ones mind with these materials. Laboratory tests show that steel and titanium alloys may be sensitive also and therefore as the proportional use of these materials increases, so we must expect stress corrosion failures to arise unless similar precautions are observed.

4. <u>The First Stage Leading to Material Selection</u>

4.1 It would not be pertinent in a paper relating to stress corrosion to make a check list of all the points a designer needs to take into account. Consideration of a few of the basic aspects are necessary to enable a proper sequence of choice to be described.

4.2 <u>Design Influence</u>

1. Consider the necessary static strength bearing in mind stiffness, fatigue behaviour, weight and cost.

2. Consider the design with relation to the anticipated dominant applied stress pattern.

3. Consider the method of manufacture :-

   Raw material - Should the part be made of sheet, plate, extrusion, forging or casting?

   Fabrication required - Bending, forming, machining.

   The method of Assembly - Welding, adhesive bonding, fasteners.

   Study the best combination of the above which fits the requirements and which can be achieved with a satisfactory low internal residual stress.

By the time this stage has been reached the designer should have formed a preference for a short list of materials which would suit the purpose. The following aspects then need to be checked :-

4. The method of flaw detection and standard of quality required needs to be established, this may cause some review of clause 4.2.3.

5. Consider the protection necessary for the type of material tentatively selected. Ensure that the design will enable peening or plating and painting to be properly performed upon all relevant surfaces. Give particular attention to positions which will cause a bi-metallic couple upon assembly. Remember that apparently enclosed and therefore safe spaces very often are not sealed and are thus susceptible to condensation.

6. Consider the manner of assembly, especially limits on fits and adjust the design accordingly so that high clamping or built-in stresses are not likely to arise.

5. <u>The Second Stage Leading to Material Choice</u>

With most of the design aspects considered and a first appreciation of the likely choice of material established, it is now possible to consider the selection in more detail.

Predominant in ones mind will be :-

1. The strength (usually it is desirable to choose the material of the lowest mechanical properties possible with the selected type which will satisfy all the other criteria).

2. Fatigue behaviour. The fatigue properties, critical crack length and crack propagation behaviour.

3. Stiffness.

4. All the properties may have to be considered for elevated temperature and corrosion resistance.

When the necessary properties of this sort have been established, then consider the stress corrosion resistance of the material.

6. <u>Final Choice</u>

Only after refining the decision by consideration of points such as these can one make a final choice. It is often possible and advantageous to allow the stress corrosion behaviour of the material to dictate the detailed differences of heat treatment or composition and enable a better stress corrosion resistance to be demonstrated by one or other of the tests.

It is felt that, although not now as essential, any improvement of stress corrosion resistance that can be induced will give a little more tolerance to accomodate variation within production processing or assembly.

There are doubtless many who will query this simple first principle approach to design and material selection, yet had these factors and the others that will come to mind been taken care of in past designs, very few of the stress corrosion failures would have occurred. During my survey as co-ordinator for the Materials and Structures Panel, the first question of one of my hosts was "Welcome, if you can tell me how to prevent stress corrosion". I demurred, he was very senior and highly qualified, and I explained that I was hoping to benefit from his experience. The impression he gave was that some new magic formula was needed to prevent stress cracking trouble.

I was taken round the establishment and shown the problem they had experienced. I believe the answer to almost all of them to be contained in the simple first principle approach given above. He knew the answer already but the knowledge had not been applied and adhered to. The only formula that was needed was the implementation of these first principles.

7. <u>The Production Influence</u>

It is apparent that lack of good house keeping and production control can quickly lead to a situation fraught with risk.

7.1 <u>Aluminium Alloys</u>

Let us consider a few aspects related to the aluninium alloys. Figure 1 (5) shows the improvement of stress corrosion resistance induced in an aluminium copper alloy material by differences of precipitation hardening times. It will be noted that for either of these times, the static mechanical properties are the same.

It would not be unreasonable, and it is known to have happened, to select a precipitation time as short as possible in the interest of production economy and still meet the required static mechanical properties. As you see, this would be wrong.

The effect has been observed and recommended by several authorities for some aluminium alloy systems and is not very far removed from the more recent forms of heat treatment applied to the aluminium zinc magnesium high strength alloys, figure 2.

The improvement with respect to stress corrosion resistance can be quite remarkable but is associated with some loss of tensile strength. It is claimed that this is counterbalanced by some improvement in toughness. There are other ways of achieving similar results, fig. 3.

It is not sufficient to carry out such treatments and relax with the assumption that other effects no longer matter. A case has been reported as follows (6).

Tube can be obtained in the aluminium 5% magnesium alloy treated at 250°C. for 24 hours to produce considerable resistance to stress corrosion. Small amounts of cold work, as for slight adjustment during installation, was shown to return the material to its original susceptibility to stress corrosion.

Improvement of stress corrosion resistance without loss of other properties has been achieved by small alloy changes. A well known example of this is the addition of silver to the aluminium zinc high strength alloys, figure 4 (7) illustrates this alloying effect. The method has found favour in Germany in particular, where it is known as AZ 74, but is seems that United States authorities are not convinced of the effectiveness of the procedure. Work is continuing in the United Kingdom.

There are, of course, less obvious aspects which may influence aluminium alloys.

Many years ago a significantly increased number of brittle stress corrosion-type failures occurred with aluminium 10% magnesium alloy castings. It was shown that this alloy normally used in the solution treated and naturally aged condition was metallurgically unstable and if subjected to a little above room temperature such as long term tropical exposure or to low temperature stoving treatments, further precipitation occurred and this was accompanied by a considerable loss of stress corrosion resistance.

After much research, BNFRA (8) showed that a small modification of the alloy, including an addition of zinc could prevent the trouble.

The author has shown that low temperature treatment of this sort can seriously alter the stress corrosion resistance of aluminium 4% Copper casting when these are in the solution treated and naturally aged condition. A condition often selected for use, as are the wrought alloys equivalents, because of the high ductillity and good crack propagation behaviour.

When one considers how small some of these changes are to the alloy in its processing perhaps one begins to see why two different countries or companies have such different operational experience with apparently similar alloys. It is suggested that such instances emphasise the uniformity at all stages of production that must be achieved to prevent unexpected happenings in service.

These instances also emphasise the importance of considering for the design, and stipulating on the drawing, the precise selection of material and its treatment, mentioned in the section 'Final Choice' of this paper.

7.2 Steels

The work of Dr. Steigerwald (9) and others has begun to show a pattern for low alloy high strength steel behaviour. It is of course, necessary to bear in mind the remarks made earlier if a steel has a good fracture toughness it is the crack initiation that is important. With steels crack initiation by stress corrosion on plane specimens would seem to be a slow process. It might also very well be that neglecting problems of hydrogen embrittlement, better protection can be achieved upon steel than upon aluminium alloys.

It would seem that as the fracture toughness and crack propagation are affected by environment it is the K.I.sc. value that should have an overriding influence. When this is done it would appear that it is not the composition which affects the threshold stress, but the tensile strength level to which the steel is heat treated.

Thus the choice of steel would seem to be dependent upon the other characteristics, such as the manufacturing demands. This appears to fit the decision by France to concentrate upon the vacuum heat treated 35 NCD 16Q, and America, such actions as the small extra silicon to produce 300m. and raise the tempering temperature.

In the author's opinion and in keeping with all the aspects to be considered regarding residual stress and their relief, a signficant feature of the choice should be the tempering temperature. Other things being reasonably equal the higher the tempering temperature of the steel the better from every view point.

With the stainless steels the situation is very similar to that of the aluminium alloys. Very small differences of composition or treatment can have quite significant influence upon the stress corrosion resistance. The effect of small differences of carbon content and the stabilisation with niobium etc., are well known to all.

The wide variation of heat treatment and effect upon stress corrosion of the stainless steel FV 520 is a good example also. Overaging is effective in the same way as for the aluminium alloys. An appreciation of steels in this category is given by N. Sien (10).

7.3 <u>Titanium</u>

The position with titanium alloys is not a clear one. So far as I know, all the environmental testing has been made with pre-cracked specimens.

Very little trouble from stress corrosion has been reported from service. Perhaps this is accounted for by the fact that the material is relatively new and rather special, and hence more detailed study of the application is made and all the appropriate production points taken care of. Also, the alloys generally used so far are of low inherent stress from heat treatment.

The principle of selection would seem to need to be the same as for alloys already discussed.

7.4 <u>Summary of Basic Material Choice</u>

It would seem that while it would be desirable to use only material completely resistant to stress corrosion phenomena this is not only likely to limit unjustifiably the efficiency of the design, but in the light of data so far available will not necessarily guarantee a safe structure. Minor effects which may upset the position seem sufficiently numerous to suggest that other snags may yet be found.

Therefore it is vital that all designs are based upon practices which enable selection of material of high toughness and crack propagation resistance and which can be manufactured with the minimum of unfavourable residual stress either internal, at the surface or induced during assembly.

It is also necessary always to take into account the grain direction. Very many materials have an acceptable stress corrosion resistance when stressed parallel to the grain.

7.5 <u>Other Properties</u>

It has been suggested earlier in the paper that the consideration of properties other than stress corrosion resistance is probably of greater importance to the design.

Increasing numbers of engineers are utilising the gradually accumulating fracture toughness data. This is really only a more precise way of expressing the old style engineers preference for a material that did not break 'short'.

In correspondence a British Aircraft engineer wrote, "I think it must be considered axiomatic that, if a component design is so susceptible to the presence of a surface defect (whether from stress corrosion or any other cause), that it fails catastrophically when the defect is small (i.e. near the limit of detectability by NDT means), then that design is a bad one however well the component behaves in the absence of such a defect".

This philosophy could be interpreted to suggest that with proper design, it is the crack initiation aspect of stress corrosion which is the important part but again only as one of the many causes. The tendency in testing today would seem to miss this aspect.

8. <u>Protective Treatments</u>

A very high proportion of the airframe structure is subjected to dynamic loading, i.e. fatigue. Most materials will corrode in aircraft environment and certainly all show much poorer fatigue behaviour when the cyclic loading and the corrosion environment occur together. Because of this, full protection by surface improvement, metallic or similar plating and organic protectives is necessary to achieve adequate service life. The same methods are likely to improve the stress corrosion resistance of the part but they are necessary from the other view point whether material is inherently stress corrosion resistant or not. It should be emphasised that this may well apply to apparently stainless materials as well as the aluminium alloys and low alloy steels.

Even in this field methods which try to improve the stress situation, peening etc., seem at this time to make a more certain contribution than organic protectives. In the past we do not seem to have taken stress corrosion much into account and little work is published of

the development of organic protection for this purpose.

This could be because :-

1. There is little knowledge of the effect of inhibition or the concentration required at the interface.

2. Very few methods or materials can so enclose the part at risk that the environment is entirely excluded and remains so during the service life.

There are some advocates who present the thesis that as materials always have cracks from some source or another and consequently the protection is cracked also, material should only be compared for stress corrosion resistance with pre-cracked specimens.

It is suggested that the operational evidence does not fully support this. It can be shown that intergranular corrosion, corrosion fatigue, and corrosion pitting can be prevented by suitable protection and design.

It would seem that material without pre-cracks, i.e. plane specimens, often take a long time to fail by stress corrosion. Further by proper protection cracking can be prevented even now without taking into account developments that are being worked on.

From all aspects the protective treatment of metals with organic protectives, or even some metallic ones, have an enormous part to play in the engineering operational life of the aircraft. Design must be such that production depts. are given a chance to make a fair job. Production depts. must appreciate their responsibilities.

In view of the relatively few materials which are fully resistant to stress corrosion at the present state of the metallurgical ark the selection of material may well have to take into account the questions. Can the material otherwise most suitable be adequately protected to resist the environment in which it will have to work? If not, then it is better to look at the next best material that is more amenable to treatment.

Perhaps I may be forgiven for emphasising that the inside of a part or fitting is usually the most difficult and thus the controlling factor in terms or selection of the best protection.

9. Non-Metallic Materials

It is perhaps not our place to refer to the behaviour of non-metallic materials. These nearly all suffer from effects similar to stress corrosion. i.e. Methylmethacrylate and other resins will craze. Some Elastomers fail by Ozone cracking.

Considerable deterioration of strength when compared with dry tests can be caused by exposure to wet environment. Structural adhesives, fibre glass laminates etc., fail much more quickly when a small static stress is applied during the exposure.

Unfortunately the development of protective methods for structural plastics has not reached a high degree of expertise. The stress-environment considerations must dominate any design.

As with metals the stress component can arise from fabrication methods or from applied service operational stresses. Methods of protection are either not practical or are not very advanced.

In these cases the corrosion aspect is even less apparent to the observer than it is with many metals. The effect can be equally devastating however.

So far non-metallic materials have not been widely used for important structures of aerospace vehicles, but more and more uses are being considered and this symposium would not be complete without drawing attention to the need to take environment in association with stress into account so that the stress corrosion story of metallic structures is not repeated.

Conclusion

The paper has attempted to put the question of choice of material and stress corrosion into perspective.

As always, good design is a compromise of the ideal monitored by availability, manufacturing methods, and not the least important, the effect of cost.

It has been suggested that in the main the design and the choice of material are interdependant and that providing the manufacturing methods and corrosion protection are taken into account the stress corrosion resistance of the material need not, indeed

probably cannot, be a controlling feature. Thus the choice of material that provides the best all round mechanical properties, at the lowest unfavourable residual stress, at an acceptable price, is the best selection.

Within this selection it is often possible to effect detailed material choice within a given type of alloy such as by minor variations of composition or heat treatment to advantage, often without an associated loss of mechanical strength.

In this way, by good design and material choice (and this does not mean by making an expensive product) despite the fact that few structural materials are inherently stress corrosion stress resistant, failures from this cause should be eliminated.

It is the author's opinion that insufficient attention to the detailed aspects of design from a material viewpoint, coupled with a lack of appreciation of the part production processes play, have led to many of the troubles of the past.

Further, nearly all these details which have to be considered are of equal importance to ensure long life in spite of corrosion fatigue, and plain corrosion and are therefore justified in any good design.

The author is grateful for the permission of the Directors of Hawker Siddeley Aviation to publish this paper. The views expressed are entirely his own, and do not necessarily reflect the views of the Company.

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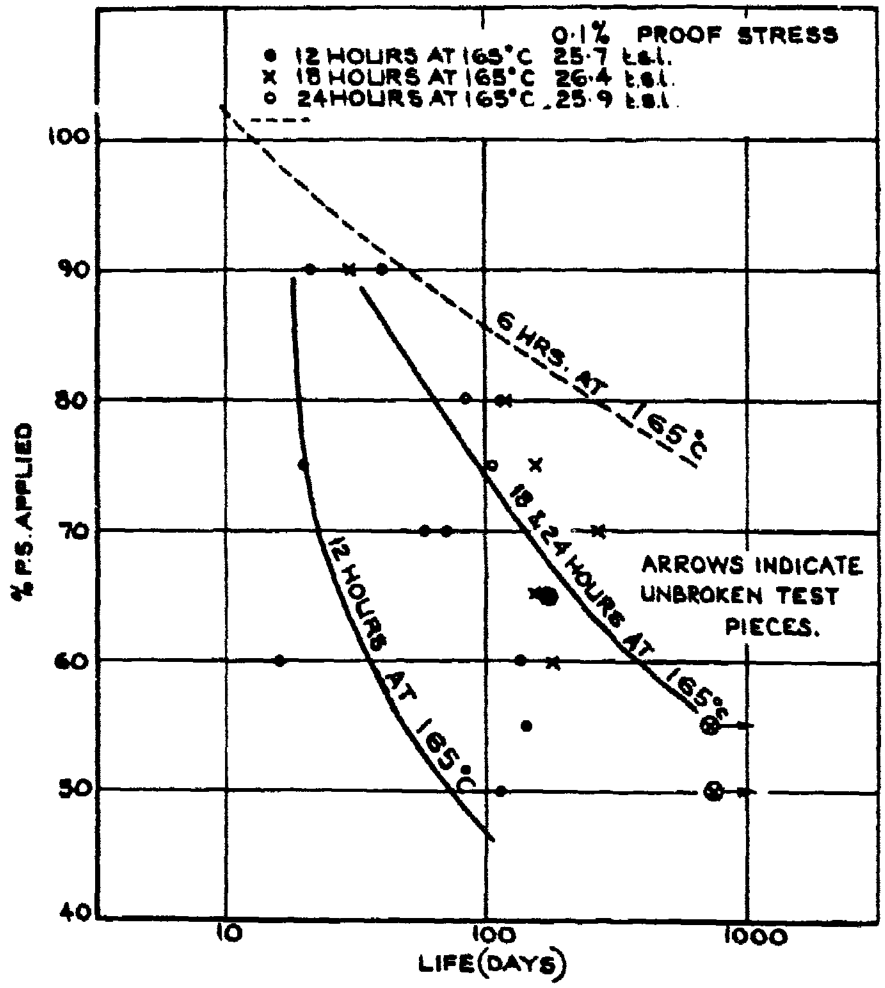

THE EFFECT OF AGEING TIME AT 165°C ON THE STRESS-CORROSION OF D.T.D. 646.

Fig. 1 The effect of different precipitation hardening times upon stress corrosion resistance. DTD.646 Al. Cu. Alloy. R.A.E. Tech. Note Met. 290.

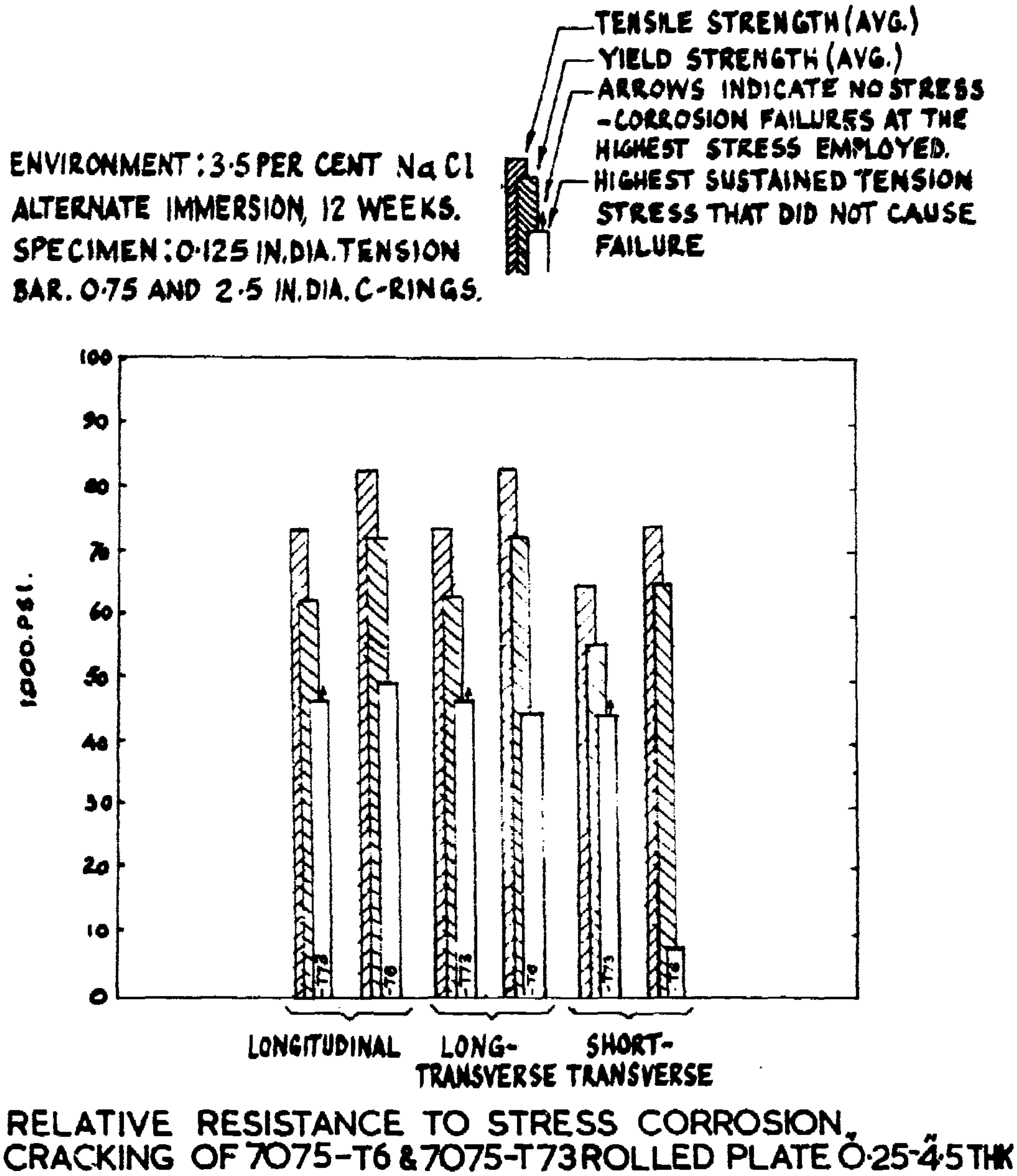

Fig. 2 The effect of T.73 heat treatment upon the stress corrosion resistance compared with T.6 heat treatment. Aluminium Co. of America.

Material and Heat treatment	Mechanical Properties: Fracture Toughness K_{IC} ST		Fracture Toughness K_{IC} LT	0.2% Proof stress ST ②		0.2% Proof stress LT		0.2% Proof stress L	Stress Corrosion Resistance ③ Life in Days: NaCl Spray @ .5% Proof stress		Atmosphere @ 90% Proof stress
7075 Type											
DTD.5094 Hot Quench @ 180°C Long Age	14.7	↑	21.7	27.7	↑	29.5	↑	31.1	89,100U ④ 100U	↑	3 specimens 209U ④
DTD.5104 Duplex Aged @ 135°C and 150°C	15.6	DECREASE	26.7 ① 22.0	28.2	DECREASE	28.4	LITTLE CHANGE	-	10,30,49	INCREASE	3 specimens 108
DTD.5024 Single Age @ 135°C	16.7		24.7	28.8		29.8	↓	29.6	6,7,11		15,16,16
2618 Type											
DTD.5084 Duplex Aged @ 200°C and 250°C	18.5	↑ INCREASE	25.5	22.1	↑ DECREASE	22.6	↑ DECREASE	23.0	361U 368U	↑ INCREASE	475U 3 specimens
DTD.731 Single Aged @ 200°C	16.5		21.4	23.4		26.0		26.2	Less than one day		307, 405 457, 581U

① Different Positions in block

② 0.1% Proof Stress Data from High Duty Alloys Ltd.

③ Stress Corrosion Data from High Duty Alloys Ltd.

④ U = Unbroken

Fig. 3 **The influence of different forms of heat treatment used to reduce stress corrosion, upon the mechanical properties of the British Al. -Zn. -Mg. Alloy Sensitivity.**

Stress-Corrosion Tests on High-Purity Al-Zn-Mg Alloy Sheet

Zn, %	Mg, %	Cu, %	Ag, %	Ageing Treatment	0.1% PS, tons/in^2	TS tons/in^2	Life, h
5.43	2.36	...	...	16 h at 135°C	22.2	24.5	0.3, 8.7, 14.7, 25.4, 26
5.32	2.36	1.01	...	16 h at 135°C	26.2	28.9	25.2, 35, 206, 238U
5.40	2.47	...	0.30	16 h at 135°C	27.3	29.8	48.5, 91, 94, 332
5.40	2.44	1.01	0.31	16 h at 135°C	28.0	31.2	335, 212U, 238U
				4.5 h at 175°C	27.6	30.6	238U, 502U

U = Specimen unbroken, test discontinued

Fig. 4 **The influence of the addition of Silver the Al. -Zn. -Mg. Alloy. Polmear, Journal Inst. Metals. 1960, Volume 89.**

THE ENVIRONMENT ENCOUNTERED IN AIRCRAFT SERVICE

by

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SUMMARY.

The paper details those factors which make up the aircraft environment, dealing first with atmospheric influences, followed by the basic aircraft materials and systems and finally those conditions imposed by the operational role of the aircraft. The importance of interaction between the individual factors in considering the total aircraft environment is stressed.

There is a brief description of work which is in hand to relate the analysis of aircraft contaminant fluids with corrosion behaviour, together with preliminary results.

Finally it is suggested that modern corrosion protection methods make it possible to largely eliminate the collective effects of the environment, provided there is an adequate understanding of the problem in design, production and maintenance.

THE ENVIRONMENT ENCOUNTERED IN AIRCRAFT SERVICE.

'Environment'. What do we mean by it? A classical English Dictionary defines it as, 'the conditions or influences under which any person or thing lives or develops'. To the designer, it often means a series of variables in terms of maximum or minimum temperatures, pressures, speeds, etc. in association with a number of well defined fluids which are to be used in the aircraft systems. It is possible, by reference to I.C.A.O. charts, tables of fluid resistance and so on, to reduce the problem to a series of hard numbers appropriate to the performance and function of the particular aircraft being considered. Were this a valid interpretation, this paper could terminate at this point with a display of the appropriate data. However, hard experience shows that the problem is by no means as simple as this. The high performance and rate of utilisation of modern aircraft means that values approaching the predicted extremes of temperature, pressure, humidity etc. can often be met in the course of a single flight, and repeated at frequent intervals, as in the case of aircraft operating between the tropics and temperate or even sub-arctic climates. Thus, what may be regarded as extremes may well be almost daily occurrences, whilst abnormal circumstances may lead to the generally accepted extremes being greatly exceeded. The consideration of the many variables as individual quantities leads one to forget that interaction between factors may modify the whole picture enormously. One has only to remember the interaction between the two factors, stress and corrosion, and the consequent effects, to appreciate that, with a multiplicity of factors involved, the situation becomes highly complex.

It is the purpose of this paper to highlight the most important factors, and to link them into a 'total aircraft environment'.

For our purpose, this total environment is conveniently divided into three phases.

First, the atmospheric environment, that is, the conditions or influences which apply in any space, whether it be in the air or on the ground, in the absence of an aircraft. Secondly, there is the basic aircraft environment, which adds those conditions which are impressed by the aircraft itself, having regard to the materials used in manufacture, and the nature of the systems carried. Lastly, we have the aircraft operational environment, which takes account of the role of the aircraft, including payload, route structure and maintenance aspects.

The Atmospheric Environment.

We have indicated that, with the increased mobility of modern aircraft, the limits of the predicted atmospheric envelope may become frequent occurrences, and it is useful here to highlight those factors and exceptions which can have significant influences. These are:-

1. Temperature.

The normally accepted maximum ambient temperature is in the region of 70°C, with allowance being made in special cases for local higher temperatures arising from engine proximity, hot air de-icing, system functioning etc. Under conditions of full sun exposure with zero wind and high ground reflection some surfaces may reach temperatures well in excess of 70°C and cases above 100°C have been recorded. Similarly the generally accepted minimum of about -55°C may under adverse conditions be exceeded by a similar amount, both on the ground and in the air.

2. Pressure.

The changes in atmospheric pressure with altitude are important, not only from a structural viewpoint, but also because of their influence in producing a breathing effect which introduces contaminant deep into imperfectly sealed cavities. This situation is further aggravated by the reduction in boiling point arising from reduced pressure (not always taken into account) which promotes more rapid evaporation of carriers such as water, often concentrating the carried contaminants at the most disadvantageous sites.

In order to appreciate the magnitude of changes in pressure it is useful to consider relative pressure under standard conditions. Thus taking pressure as being equal to 1 at sea level, this drops to ½ at about 18,000 ft., ¼ at 34,000 ft. and ⅛ at 48,000 ft.

3. Humidity/Water.

In most areas water exists in abundance in any of three forms vapour, liquid or solid, and, under the right conditions, will rapidly change from one form to another. The presence of water is the real enemy in aircraft corrosion in that, being abundantly available, and having exceptional solvent ability, it acts as an efficient carrier for dissolved or suspended contaminants and forms the basis of all good electrolytes.

4. <u>Contaminating Influences.</u>

These are largely dependant on geographical location and can arise from two types of source. Those which are naturally occurring, and those which are impressed upon the site by the presence of man. The former category includes those influences which arise from proximity to the sea or local terrain, airborne matter such as, rain, hail, ice, dust and dissolved chemicals, and ozone at higher altitudes. Among other naturally occurring phenomena which have an important bearing on the degradation of materials are the spores of micro-organisms and radiations of varying wavelengths, including infra-red, visible light, ultra voilet and cosmic rays. Man complicates these influences by the addition of animal secretions and domestic and industrial effluents.

<u>The Basic Aircraft Environment.</u>

The aircraft modifies the existing atmospheric environment by the addition of the wide variety of materials used in its construction and carried in its systems. Those influences incorporated during build include, apart from the basic constructional materials, stresses from various causes and residues from handling and treatments. The stresses contained within the airframe arise essentially from two counts. viz. persistent stress levels within the material stock from which it was made, or induced by the part fabrication process and, secondly, stresses induced by the assembly practice and arising from eccentricities, mismatch of parts, force fitting or interference fit fasteners etc. Additive to this is the basic stress within the airframe inherent from its geometry and mass. Operational stresses add a third contribution and these depend on the type and role of the aircraft. It should be noted that it is now usual for the modern commercial airframe to achieve annual utilisations between 3 and 4,000 hours so that operational stresses are superimposed on those inbuilt for some 40% of the aircrafts' total calendar life. Thus, since it is the aim of good modern design to control and/or reduce to a minimum all the non-useful stresses enumerated above, the stresses arising from operation are tending to become the critical factors in stress corrosion phenomena. This is in contradiction to past thinking where, due to relatively low aircraft utilisation, the persistent inbuilt stress was considered to be the significant stress environment.

Although it would appear at first sight that there is little hazard from trace residues resulting from handling and treatment of components some factors merit careful consideration. Salt bath treatment of rivets often takes place shortly before use and, if subsequent rinsing is not efficient, then salts are carried directly into joint areas. Similarly, the fitter who wipes his hand over a close tolerance bolt to ensure that it, 'feels' clean may introduce a dangerous contaminant into a very vulnerable area.

Freezing of components as an aid to shrink fitting often results in frost formation which may introduce water in to the heart of a vital joint.

1. <u>Fuel.</u>

Although it is the fluid used in the greatest quantities by the aircraft, fuel poses very few problems in the corrosion context. Due to its flammable nature, all possible steps are taken to prevent spillage, and, because of rigid engine requirements, the impurity level is generally low. Thus problems from this cause only normally arise from local accidental spillage during refuelling and maintenance, and because of its relatively inoccuous nature, only certain non-metallic materials are likely to be affected.

Within the interior of the fuel cells, however, a particular problem can arise due to the development of fungal growths. The spores of micro-organisms, particularly the fungus Cladosporium resinae are present in many fuel supplies and, given the right conditions within the fuel cell, can germinate and grow. Large scale efforts have been made in the design and maintenance of aircraft to combat the conditions which lead to fungal development, but there is still a world wide problem. This arises mainly from the fact that water can exist in the fuel tanks in globular form. Although in the normal concept of drainage these globules should be regularly removed, they do in practice, attach to the tank interior and do not gravitate towards the drain points. Given these conditions and temperatures in the region of 30°C, all the ingredients for growth are present. Apart from the danger of filter blockage, which falls outside the scope of the present paper, metabolic products of the organisms are extremely corrosive and, at any gap in protective coating, or where protective coatings are penetrated, severe intergranular corrosion can occur in very short periods. Control is at present largely achieved by the use of biocides, but this approach must be used with caution since, with injudicious use, other equally large problems may arise either from overdosing, or the effects of thermal decomposition products within the engines.

2. Hydraulic Fluids.

Here again design integrity demands that these fluids are well contained, but, because hydraulics provide a service throughout most areas of any aircraft, minor spillages or failures can in this case result in widespread contamination. The problem is aggravated by any neglect in design to provide adequate containment or disposal of spillage, and by poor husbandry on the part of the user. In the past, hydraulic fluids based on mineral oils were largely inhibitive to corrosion and inert to most materials. The more modern phosphate ester fluids, however, have very severe effects on many organic materials thus exposing metal to attack, and may, on hydrolysis or thermal degradarion, more readily give rise to corrosive substances.

3. Domestic Water, Galleys, Toilets.

These facilities form a collective group in that they all serve the personal needs of the crew and passengers. Domestic water, if spilled, will act as a carrier, as already discussed, but in this case, if chlorinated, will also become dangerous in its own right. Bad husbandry again poses the main problem as the most appalling misuse of concentrated chlorination compounds occurs on occasions, due either to bad premixing or the use of massive overdoses during sterilisation of the system. In this case, apart from disastrous results within the water system, the effects of any spillage will be much more serious.

The immediate problems of the 'kitchen' are well understood and steps are taken to contain major spillage of food and liquid. However, these efforts are not always sufficient or successful, and one aspect requires serious consideration. Spillage of food and residues, fruit juices, etc., frequently moistened by steam from coffee dispensers or condensation, provide ideal breeding grounds for bacteria which, by their metabolism, can produce corrosive by-products to attack structure.

Toilets and toilet waste areas are again obvious breeding grounds for bacteria but the work we have done to date indicates that the strong disinfectant measures which are taken in these areas are largely effective in controlling the growth or micro-organisms. The contents of the toilet systems themselves, however, present a serious corrosion hazard if major spillage occurs.

4. Electrical Systems.

Electrical systems contribute to the overall corrosion problem in two ways. The first, and most obvious effect is from batteries, which may be lead-acid or alkaline. Apart from spillage of the electrolyte, batteries form a significant source of corrosive fumes at the high charging and discharging rates which can occur in operation.

Not so obvious is the contribution made by the electrical system itself, by introducing potential differences within and between metallic materials either through the use of the structure as an earth return or conducting path, or by induced current from power lines, navigation and radio aids, aerials etc. Thus local heating may occur, and impressed voltages form corrosive cells where electrolytes are present. Too little study has been devoted to these effects, but an illustration from industry provides a sober warning. A very large processing tank, correctly installed in an aircraft manufacturing works failed after some two weeks use when the bottom failed through corrosion. There was no reason evident until a person, totally unconnected with the problem observed in passing, 'I hope you have not damaged the main power feeder cable which passes underneath that tank.' The tank was moved and the problem solved.

5. Air Systems.

The cabin conditioning system produces little environmental hazard in its own right except by acting as a collecting and redistributing system for the impurities in the outside air which it draws either directly or via the engine compressor stages. It also collects passenger respiratory products and tobacco tar which it deposits at preferential points either at exits, or within the system.

6. Fire Extinguishers.

The automatic fire extinguishing systems are mostly of the inert gas type which are relatively innocuous. Hand extinguishers carried aboard the aircraft are usually carbon dioxide or water and, with intelligent use, do not constitute a serious problem. The main corrosion hazard, therefore, arises from the ground incident or accident, when conventional fire fighting services use materials varying from dry powders to foams, or even soda-acid mixtures. Some of these methods leave absorbent residues, whilst others are directly corrosive in their own right. Casual, incomplete cleaning can, therefore, be dangerous, and there is sometimes a tendency to 'neutralise' residues with chemicals which are as dangerous as the orignal contamination.

7. Anti-icing and Defrosting Systems.

Cold weather operation involves the use of large quantities of de-frosting or anti-icing fluids and compounds. Most are based on alcohol mixtures in liquid or gel form, the latter sometimes applied by hot spray from special dispensers. Little immediate deleterious effect arises from this cause, but little is known about long term effects.

8. Heating/Cooling Effects.

Rapid thermal changes are the remaining significant constituent of the basic aircraft environment. Some of these arise from systems operation, already discussed, but it must be remembered that, where heating or cooling units are designed for operation in flight, exaggerated effects will be obtained from functioning on the ground. Abnormally high temperatures, with consequent high thermal stresses may occur with heating systems, such as leading edge hot air de-icing, whilst intercoolers may act as condensing towers under conditions of high humidity, and will also add their quota of thermal stresses.

Supersonic flight gives rise to substantial kinetic heating of forward facing and external surfaces, with maximum temperatures in the region of 120°C to 150°C being reached on current aircraft. Apart from the obvious need for materials of higher thermal stability and higher strength at elevated temperatures, this leads to greater thermal stresses, both transient and persistent, and increasing the risk of concentration of contaminants by rapid evaporation of carrier liquids.

The advent of vertical take-off aircraft leads to similar problems during the extreme take-off and landing stages, due to ground reflection of heat and engine exhaust.

Aircraft Operational Environment.

1. Payload.

In passenger carrying aircraft a large number of people are concentrated into the minimum possible space. High rates of air change ensure their relative comfort, but respiratory and perspiration products together with tobacco tar may condense and become concentrated within the cabin or system ducting. On entering the aircraft passengers will bring with them traces of soil, water possibly de-icing fluids or salt, and an infinite variety of potential contaminants in hand luggage.

The carriage of freight brings many more problems, which are on the increase as the variety and volume of freight expand. Almost anything from fresh fruit through chemicals to livestock is carried, and sooner or later, all will spill or be spilled. Once again, awareness of the problem on the part of designers and operators can minimise the problem but the penalties of failure can be grave. Drawing an example from the writer's experience, fully charged lead-acid batteries packed in cardboard containers were loaded in an inverted position in the forward pressurised freight hold of an airliner. The fault was noticed at the first stop, some 700 miles away, and local corrective action taken by 'neutralising' with a gelled caustic soda formulation, intended for, and better suited to, the cleaning of domestic ovens. The aircraft flew for a further three days before an inspection at main base, when it was found that heavy contamination had spread throughout the length of the aircraft, even re-entering the aircraft from outside via cooling air ducts and control surface apertures. The subsequent down time and repair bill were both enormous. The lessons to be learned are for the designer to make it both possible and practicable for such spillage to be effectively removed without being trapped in inaccessible areas, and for operators to recognise the dangers of spillage, and to make all maintenance personnel aware of the first aid action needed to minimise, and not aggravate, the danger.

2. Route Structure or Role.

This plays a profound part in the airframe environmental exposure pattern. Short haul aircraft involve a large throughput of passengers, frequent ground handling and temperature excursions (often too short for conditions to stabilise) and, with today's high mobility, may even move from one climate pattern to another in the course of a normal tour. Hence transient thermal stresses tend to be both high and repeated, whilst condensation problems and contamination by passengers and terrain are frequent. The long haul aeroplane spends sufficiently long periods at high altitudes to reach stable cold conditions and during normal turn-round times, water in fuselage bilges or fuel cells will remain frozen, making draining procedures ineffective, and allowing a successive build up with each flight. Measurements taken of wing integral fuel tank structural temperatures during a 7½ hour flight, cruising at 38,000 ft, showed that stabilisation occurred at between -20°C and -30°C only after approximately six hours. Ice would persist in the tanks for equal or longer times after landing.

A word of caution is required against concluding that aircraft can conveniently be alotted into definite patterns according to their routes and usage. Experience has shown that three European operators, using the same type of aircraft over essentially the same routes and with pooled maintenance each developed their own distinctive pattern of corrosion behaviour. The differences, which could only have arisen from the individual fitting out of the aircraft, and the different modes of operation, were never satisfactorily explained. The only conclusion which can be drawn is that each operator will superimpose his own unique pattern upon the general environmental conditions.

Since the end of the second world war, the utilisation of individual aircraft has increased dramatically from a few hundred hours per year to 3 - 4,000, and, at the same time, the range and scope of aircraft has reached out from almost local operations to the far corners of the globe, and into places previously not dreamed of. Our thinking must keep pace with this expansion, maintaining reliability under arduous conditions.

3. Maintenance.

Many examples have already been quoted of the importance of action by the designer in making effective maintenance possible, and by the operator in carrying out this maintenance to a high standard. Good housekeeping is an extremely important factor in preventing trouble and we in the manufacturing industry must take our share of the responsibility for passing on all available information. It is up to the operator to ensure that materials and methods used in cleaning, paint stripping and re-protection are the right ones, and are used in a safe fashion, and up to the manufacturer to provide adequate guidance.

Design concept must recognise that the standard of available facilities (and this includes personnel) will vary enormously, whilst the operator on his part must realise that what held for a previous generation of aircraft may not necessarily hold for new and more sophisticated designs. Local practices and customs too, play a part, and whilst, on the credit side, the local man knows his own country and its peculiar problems best, on the other hand this should not be allowed to stop progress or the introduction of new and more effective methods. The pattern of maintenance is changing, with more and more operators using the 'progressive overhaul' system. This brings new problems in its train, as it is becoming a rare occasion for an aircraft to be grounded for more than a very few days for maintenance. Consequently, operations such as painting, which, because of the use of more aggressive fluids and higher demands on performance, become increasingly important and time consuming, have to take place in shorter and shorter periods. It is, therefore, vital that such processes are properly planned into the schedules and not left to the unfortunate hangar foreman to fit in where possible.

Recommendations.

We have seen that a great many factors contribute to the total aircraft environment, and that it is not valid to consider the effects of the individual factors in isolation. It follows from this that laboratory tests, using relatively simple corrodants, high stresses and short timescales can, at best, only give qualitative, sorting data. It is our view that much more needs to be done to obtain real information on the composition and properties of the fluids which accumulate in, or wash over, aircraft structure. Under the sponsorship of the Ministry of Technology, we are carrying out a programme of sampling these fluids, and correlating their chemical analysis with corrosion and stress corrosion behaviour. This, coupled with a regular examination of the aircraft, and incorporating microbiological sampling of selected areas, will enable a much more realistic appreciation to be made of the problem. The work is in an early stage as yet, but already some patterns are emerging as shown in Tables 1 and 2. Cases are being found (see samples 5 and 10) where liquids which, on simple chemical analysis, appear to be relatively innocuous cause severe and rapid corrosion whilst other fluids, high in chloride content, are found to cause little corrosion (see sample 6). Deeper analyses are to be made to explain these anomalies, and it is hoped that a realistic grading of individual constituents may give designers information which can be used to isolate or combat any particularly dangerous substances or areas or, equally important, take advantage of any beneficial factors.

Despite the complexity of the problem, and the many inadequately explored factors, it is our experience that modern corrosion protection materials and methods, which have made enormous advances over the last decade, make it possible to largely eliminate the collective effects of this environment, provided they are properly applied and maintained. Nevertheless, improvements, and a better understanding of the situation are needed if full advantage is to be taken of these advances.

In conclusion, therefore, we would summarise those aspects which we consider require particular attention as follows:-

1. It must be recognised that an aircraft designed for a specific customer and operational role may change hands a number of times during its useful life, and design for world wide usage should, therefore, be the aim.

2. Design must take account of the problems of production and maintenance, and conversely operators must appreciate design philosophy. Quality Control must ensure that what was aimed at in design is achieved in production and service.

3. Airframes should be designed and manufactured to ensure rapid clearance of condensates and contaminants into drain areas which automatically drain free and are accessible to inspection. Drain paths are useless if they are too small and become blocked in service.

4. Areas of particular contamination, such as battery bays cannot be treated in isolation to a higher standard than surrounding areas, as aircraft movements and carrier liquids will spread the contamination. In effect, this means that the general level of protection should be capable of dealing with the most aggressive contaminant.

5. The most vulnerable areas of most airframes are joints and close tolerance/working surfaces, where the full standard of protection is often omitted for various reasons. Pre-painting with stoved chemical resistant primer and finish, supplemented by an adequate wet assembly technique, is capable of virtually eliminating corrosion problems from most static joints. However, there is still room for considerable improvement in protecting close tolerance and/or working surfaces in corrodible materials. The lessons learned in the protection of magnesium alloys could well be applied in this instance, that is, if it is not possible to properly protect a part or surface made from corrodible material, then either the material or the design must be changed.

Table I.

Sample No.	Aircraft	Location	pH	Chloride g/l	Sulphate g/l	Iron	Chromium	L85 (3 Weeks) Corrosion Test Weight Loss (mg.)
1	Turboprop	Under Flap Motor.	8.3	0.19	Trace	Absent	Absent	1.4
2	Turboprop	Fwd. of Rear Freight Bay.	8.4	1.90	Present	Absent	Present	5.4
3	Turboprop	Centre Fuselage.	7.2	0.13	Trace	Absent	Absent	2.0
4	Turboprop	Under Flap Motor.	8.0	1.06	Present	Absent	Absent	1.3
5	Turboprop	Centre Fuselage.	6.0	0.15	Trace	Absent	Absent	2.7
6	Turbojet	Battery Bay.	8.2	1.09	Trace	Present	Absent	0.3
7	Turbojet	Under Toilet.	6.2.	0.11	Trace	Absent	13 ppm.	0.1
8	Turboprop	Centre Fuselage.	8.3	0.64	Trace	Absent	Absent	4.5
9	Turboprop	Under Flap Motor.	7.3	0.75	Trace	Absent	Absent	0.5
10	Turboprop	Nose Radome Area.	7.5	0.05	Absent	Absent	Absent	3.8
11	Turboprop	Centre Fuselage.	5.5	1.18	Absent	Absent	Absent	4.9
12	Turboprop	Centre Fuselage.	7.3	0.43	Trace	Absent	Absent	1.6
13	Turboprop	Centre Fuselage.	8.5	0.58	Trace	Absent	Absent	3.5
14	Turboprop	Rear Fuselage.	7.6	0.11	Trace	Absent	Absent	1.8
15	Turboprop	Front Fuselage.	7.9	0.10	Absent	Present	Absent	1.4
16	Turboprop	Centre Fuselage.	7.9	0.29	Trace	Present	Absent	3.3

Table 2.

Control Corrosion Test.

Chloride g/l	% Sulphate	Weight Loss 1 week (mg)	Weight Loss 3 weeks (mg)
0.1	Nil	2.3	6.0
1.0	Nil	2.7	7.1
5.0	Nil	2.4	6.0
10.0	Nil	2.3	7.1
20.0	Nil	2.7	8.5
Control - Distilled Water		1.0	3.1

THE INFLUENCE OF SURFACE AND PROTECTIVE TREATMENTS ON STRESS CORROSION BEHAVIOUR

by

H. G. Cole

The influence of surface and protective treatments on stress corrosion behaviour

by H. G. Cole

Summary

Data are presented on the influence of surface and protective treatments on the stress corrosion behaviour of magnesium alloys, aluminium alloys, titanium alloys and low alloy and stainless steels. Compressive surface stresses introduced by peening are beneficial on all metals. Sacrificial metal protective coatings are beneficial on all metals except low alloy steels in a state highly sensitive to hydrogen. Conversion coatings give little protection alone or may be deleterious but in conjunction with high grade paint schemes give fair protection. Brief recommendations are added for the application of existing knowledge and on directions for future work.

1. Introduction

This paper concerns the influence on stress corrosion behaviour of surface mechanical treatments, such as shot peening, and of protective treatments such as metal coatings, conversion coatings and paints. These treatments are applied chiefly for reasons not connected with stress corrosion, but it is clearly important to know how effective they are in stopping stress corrosion cracking.

2. Magnesium alloys

2.1 Introduction

Although a few stress corrosion failures of cast magnesium alloys and of wrought Mg-Zn-Zr alloy have been reported, the problem of stress corrosion in magnesium alloys is almost wholly confined to the older Mg-Al-Zn alloys in wrought form. The Mg-6Al-1Zn alloy is more susceptible than Mg-3Al-1Zn. Corrodents causing stress corrosion are distilled water and chloride solutions.

2.2 Surface mechanical treatments. Shot peening has been advocated as a method of preventing SCC(1) but the author does not know whether it has ever been used. If metallic shot were used, a cleaning treatment such as fluoride anodising(2) would be necessary to remove any surface contamination.

2.3 Metallic coatings. Stress corrosion of magnesium alloys can be stopped by cathodic protection, and a sacrificial cladding of Mg-Al-Zn alloys with Mg-1Mn alloy, not itself susceptible, has been shown to be effective(3), but the author again does not know whether this method has ever been used in practice.

2.4 Conversion coatings. No information has been found on the effect of chromate treatments. As, however, a mixture of chromate and chloride has been used for stress corrosion testing, it may be assumed that chromate films alone will not protect against stress corrosion failure.

2.5 Paints. Paint has been reported to delay failure(4).

2.6 Conclusions. Stress corrosion is not a problem with the current Mg-Zn-Zr alloys or with the cast forms of the older lower strength Mg-Al-Zn alloys. The susceptible wrought forms of the older Mg-Al-Zn alloys can be treated by shot peening or by cladding with a sacrificial Mg-1Mn coating. It is the author's opinion that the conventional protective treatment of chromate treatment plus organic finishes would be effective on a material of low sensitivity to SCC, e.g. Mg-3Al-1Zn stress relieved, but not on a more susceptible material, e.g. Mg-6Al-1Zn hard rolled.

3. Aluminium alloys

3.1 Introduction. All the stronger alloys are susceptible to stress corrosion by chloride solutions, and the more susceptible materials by an environment as mild as a damp atmosphere.

Extensive investigations by Alcoa, LTV Inc., Boeing and the George C. Marshall Space Flight Centre have been summarised in reference (5). Table I is reproduced from this publication.

3.2 Surface mechanical treatments. Peening is widely used to reduce the susceptibility of aluminium alloy components to stress corrosion. Waller(6) considers it one of the most important steps. Table I line 1 shows the effectiveness of peening on several alloys especially under natural exposure conditions. Similar results have been obtained in Germany(7) where the effect is attributed to surface deformation rather than to compressive stresses.

Hawkes(8) has determined the stress corrosion thresh-hold of short transverse Al-Zn-Mg specimens under a salt water drip. The specimens were prepared to give three states of

initial surface stress before peening and the peening was carried out with rust-free steel shot to Almen intensity 0.008A2. The results, given in Table II, show that peening gave a high stress corrosion thresh-hold even on material which had tensile stresses in the surface before peening. The balancing tensile stresses within the peened metal were low in magnitude.

Peening can be carried out with either steel shot or glass beads. Steel shot is preferred for obtaining a good depth of compressed layer(5), but iron embedded in the surface must be removed; Waller(6) immerses peened components in 25% nitric acid solution followed by a ferrocyanide test for absence of iron on the surface. Stainless steel shot is being investigated(5) as it causes less contamination.

Peening is carried out with shot of diameter about 0.25mm (0.01 in) to an intensity of 0.005A2 on the Almen scale. The depth of the compressive layer is roughly equal to the diameter of the shot and maximum compressive stress just below the surface is about 24 hbar (35,000 lb/in^2).

3.3 Metallic coatings, cladding. The stress corrosion of aluminium alloys can be prevented by cathodic polarisation, and the sacrificial coatings of pure Al or Al-Zn alloy applied to sheet to protect against corrosion give also a high degree of protection against stress corrosion. This property of cladding seems to be so widely accepted that the author has been unable to find a reference to published work, though no doubt many metal fabricators and aircraft manufacturers have unpublished reports in their archives. Stress corrosion can occur in sheet when the cladding is substantially corroded away, (Figure 1).

3.4 Metallic coatings, sprayed. Cladding is applied only to sheet. On forgings and extrusions comparable protection against stress corrosion is given by sprayed sacrificial coatings.

Line 4 of Table I shows the high degree of protection given by sprayed Al-1Zn to five U.S. alloys. The results of a detailed investigation(9) of two British compositions, HE15 Al-Cu-Mg alloy and DTD.683 Al-Zn-Mg alloy, exposed to industrial and marine atmospheres are given in Tables III and IV. On the Al-Cu-Mg material all the sprayed coatings increased the life by a factor of at least 10 even when a 1.5mm gap was present in the coating. On the Al-Zn-Mg material all the sprayed coatings gave protection, and a high degree of protection was given by coatings of low copper content (<0.01%), though protection across gaps was less efficient.

Similar results have been obtained(10) on medium strength weldable Al-4½Zn-2¼Mg alloy sprayed with zinc. In a salt spray laboratory test, specimens stressed at 25hbar (36,000 lb/in^2) failed in 5 weeks when untreated but zinc sprayed specimens were unbroken after two years.

It is important that the surface of the component be adequately roughened prior to metal spraying. Coarse alumina grit is preferred(11). A process specification for the protection of aluminium alloys by metal spraying is being prepared in the U.K.

The use of metal spraying is limited in practice by the need to roughen the surface, by the thickness of the coating (> 50 microns), by the difficulty of blasting and spraying into re-entrant angles, and by lack of confidence in consistent long term good adhesion. Nevertheless the method is highly effective and could well be quite widely used.

3.5 Metallic coatings, electroplated. Line 8 of Table I shows that electro deposited zinc is a very effective protective against stress corrosion. Although satisfactory electroplating onto aluminium alloys is not easy, the method might find its uses in special circumstances.

3.6 Metallic coatings, zinc rich paint. Line 10 of Table I shows that zinc rich paint is effective in protecting against stress corrosion, and this material also might find its uses in service.

3.7 Conversion coatings, anodising. Line 7 of Table I refers to sulphuric acid anodising and shows that anodising accelerates stress corrosion cracking unless the anodic film is sealed with chromate. Some earlier work on British materials(12), summarised in Table V, showed that anodising by the chromic acid process also has some deleterious effect. The harmful effect of anodising is ascribed to small-anode-large-cathode effects at cracks in the anodic film. The Eloxal process, on the other hand, has been found to delay failure(13).

3.8 Conversion coatings. Table I lines 6 and 9 show that the protective efficiency of chromate filming and other conversion coatings is variable. Reductions in stress corrosion life can be caused, perhaps because the concentration of inhibitor at the metal surface is not large enough to maintain overall protection so that small anodic areas can develop where protection breaks down.

3.9 Organic protection. The experimental evidence on the effectiveness of paint schemes is somewhat conflicting. In a test programme(9) run parallel to the tests on sprayed metal coatings, Tables III and IV, an aircraft paint scheme applied after either anodising or chromate filming gave protection varying from none to moderate. Very variable results, some poor, were given in the tests recorded in lines 2, 3 and 5 of Table I. On the other hand, a British expert(14) believes that modern paint schemes are very effective on modern materials of relatively low susceptibility to stress corrosion, and that stress corrosion remains a problem only on areas which cannot be painted. Waller(6) also appears to believe in the effectiveness of paint.

An explanation for some of the discordance can be deduced from the results in Table V. The third material mentioned in this Table clearly had a very low resistance to stress

corrosion, and paint (of the type used in 1950) was ineffective at protecting it. The first and second materials in the table, however, had greater intrinsic resistance, and to these paint did give a measure of protection. The author believes that the effectiveness of paint in protecting against any form of corrosion is proportional to the intrinsic resistance of, the metal being protected. Paint must be applied to aircraft aluminium alloys to protect them against general corrosion, and this paint will also protect against stress corrosion provided that the metal is not too susceptible.

3.10 Conclusions

High strength aluminium alloys can be protected against stress corrosion by the introduction of surface compressive stresses and by sacrificial metal coatings. Of these methods, only the protection of sheet by cladding is as widely used as it should be. Conversion coatings alone are ineffective or deleterious. When followed by a good paint scheme, however, protection is good provided the intrinsic susceptibility of the metal to stress corrosion is low.

4. Titanium alloys

4.1 Introduction. Titanium alloys can suffer stress corrosion cracking in chlorinated hydrocarbons, by salt at elevated temperatures, and at room temperature by salt water and chemical agents(15).

Stress corrosion at room temperature probably occurs by a hydrogen embrittlement mechanism, and it may be assumed that the higher the initial content of hydrogen, the more susceptible the metal will be to further hydrogen introduced by a corrosion reaction. It is therefore important that the initial content be kept low. This is achieved without difficulty by the choice of suitable methods of manufacture and pickling, and of electroplating when this is carried out.

4.2 Surface mechanical treatments. Peening with silica and prolonged vibration with alumina triangles have been found effective in preventing hot salt stress corrosion(16), the effect being ascribed to compressive stresses in the surface(17).

Aqueous salt solution causes stress corrosion in practice only if a crack is already present and no experiments seem to have been made on the effect of peening; it may be assumed that peening would, however, delay or prevent the formation of an initial fatigue crack and would delay propagation by stress corrosion until the crack had penetrated the compressed surface layer.

4.3 Metallic coatings. Zinc applied by spraying or dipping has been found effective against hot salt stress corrosion; aluminium(18) and nickel(16,17,18) were also effective when free from pores. As it has been found that hot salt stress corrosion is stopped by anodic polarisation(19), the reason why a porous coating of zinc is effective whereas a porous coating of aluminium is not, is not clear.

No references have been found to the effect of metallic coatings, or of anodic or cathodic polarisation, on the propagation of cracks in aqueous salt solution.

4.4 Conversion coatings. Anodic coatings were reported(20) to reduce hot salt attack, but later work reports no effect(21). No information on the effect of anodising under aqueous conditions has been found.

4.5 Organic protection. Polyimide coatings have been found to prevent hot salt attack for about 1000 hr. at 315°C(16) but the coatings tended then to degrade and peel off. No information has been found on the effectiveness of paints, e.g. those with leachable chromates, on crack propagation in aqueous salt solutions.

4.6 Conclusions. Stress corrosion cracking in chlorinated hydrocarbon degreasants can be prevented by suitable additions to the degreasant(15) and no treatments of the metal are needed. If hot salt attack is in fact a practical problem, then metallic coatings could probably be used as protectives, though further work is needed including an investigation of the mechanism of protection in relation to the mechanism of the attack. Little work seems to have been done on protection against crack propagation in aqueous salt solutions; experiments on wet crack propagation under impressed potential or in the presence of inhibitors would suggest methods of protection.

5. Non-stainless steels

5.1 Introduction. Steels of tensile strength rexceeding 140hbar (203,000 lb/in^2) can suffer stress corrosion failure in marine and industrial environments probably by a hydrogen embrittlement mechanism. It is difficult to distinguish the failure from sustained load failure (static fatigue) caused by initial hydrogen embrittlement from cleaning and protective treatments. Although much effort has been devoted to avoiding this initial embrittlement, little work has been done on the effectiveness of coatings in protecting against stress corrosion in service. There seem to be two reasons for this: (a) Doubt as to whether initial embrittlement by the cleaning and protective process could be completely avoided and so play no complicating role in the subsequent stress corrosion exposure. In the case of metallic coatings, this doubt can now be eliminated by the use of vacuum deposition. (b) The impact of fracture mechanics which has led experimenters to see more reward in fundamental work.

These two considerations have led to a greater emphasis or fundamental stress corrosion work under impressed potentials and to a neglect of practical exposure tests.

5.2 Surface mechanical treatments. Shot peening is as effective on steel as it is on other metals. Results published by Davies(22) are reproduced in Table VI. Similar results have been reported by Barer(23) for maraging steel.

5.3 Metallic coatings. The standard protective treatment for steel parts in aircraft is a coating of cadmium deposited usually by electro-deposition and more recently by vacuum evaporation. A large number of papers have been published on how to reduce or avoid hydrogen embrittlement by electroplating but very few on whether cadmium protects high strength steel against stress corrosion.

Sink(24) carried out salt spray tests on stressed rings of AISI 4340 at a tensile strength of 185hbar (270,000 lb/in^2.) Cadmium plate prolonged the life to failure of rings stressed at less than 117hbar (170,000 lb/in^2) but caused more rapid failure at higher stresses.

Some recent unpublished results from work still in progress in the U.K. are given in Table VII(25). All plated specimens were given a baking treatment which enabled them to withstand the highest test stress in a laboratory atmosphere. The tentative conclusions can be drawn that cadmium plating sometimes gives a high degree of protection but sometimes appears to accelerate failure.

Coatings other than cadmium were tested by Phelps and Loginow(26) on 5Cr-Mo-V steel. Sprayed aluminium gave excellent protection and nickel-diffused-cadmium was very good. Nickel and chromium gave poor protection. Chromising also appears deleterious (Table VII).

5.4 Paints. The British work(25) included tests on phosphate treatment plus an aircraft paint scheme(Table VII). The number of specimens was small but the results consistently good. Phelps(26) obtained good results from high temperature paints pigmented with zinc or aluminium and good results have been obtained from aircraft paint schemes(27).

5.5 Conclusions. Too little work has been done on the protection of very strong steels against stress corrosion, the problem having become confused with plating embrittlement and with fracture mechanics. Peening seems to be undoubtedly beneficial, and phosphate treatment with an aircraft grade paint scheme also seems to give good results, at least while intact. Cadmium coatings are, however, suspect.

Nevertheless,the aircraft designer continues to call for cadmium, and, provided the process of depositing the cadmium does not cause embrittlement, appears to obtain good results in service.

6. Stainless steels

5.1 Introduction. Denhard(28) reviewed the stress corrosion properties of structural stainless steels and found that the number of reported stress corrosion failures in aircraft applications was few. Only the chromium martensitic type were appreciably susceptible to stress corrosion, and then only when tempered at a relatively low temperature.

The danger of stress corrosion is therefore usually eliminated by suitable choice of alloy and/or heat treatment. The following protective treatments are available if required:

Peening. Peening is effective. To avoid contamination of the surface with iron, it should be carried out with stainless steel shot or with non-metallic shot such as glass beads.

Metallic coatings. Sacrificial coatings such as sprayed aluminium, cadmium and zinc are effective(28):in contrast to low alloy steels, stainless steels are not susceptible to the small quantities of hydrogen liberated by the sacrificial action. Nickel and chromium do not protect.

Organic coatings. Aircraft paint schemes can be expected to give a fair measure of protection.

Some unpublished results of salt spray stress corrosion tests on FV520(S) precipitation hardening 15/5 steel(29) showed the importance of a clean surface; residual heat treatment scale promoted stress corrosion whereas thorough cleaning, especially by electro-chemical polishing, gave a high resistance to failure.

7. Conclusions

Surface and protective treatments are available for all aircraft metals which delay or prevent failure by stress corrosion. The introduction of compressive stresses into the surface by peening or other methods is effective on all metals. Sacrificial metal coatings are effective on all metals except low alloy steels in a state highly sensitive to hydrogen. Conversion coatings alone give little protection or may be deleterious, but in conjunction with high grade paint schemes give fair protection.

8. Recommendations

(a) The aircraft designer should be willing to use treatments (peening, metal spraying) which are especially effective against stress corrosion.

(b) The metallurgist must continue to reduce the intrinsic susceptibility of strong metals to stress corrosion. He should not place too much reliance on protectives because they may be difficult to apply to some components and may become damaged in service.

(c) More work is required to establish the conditions under which sacrificial metal coatings protect or do not protect very strong low alloy steels against stress corrosion.

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TABLE I. EFFECTIVENESS OF SURFACE TREATMENTS ON ALUMINIUM ALLOYS

(Ratios of median lives of the specimens with the protective systems to the median lives of unprotected specimens.)

	Protective System	2014-T651					7079-T651 T6					Alternate Immersion		
		Alternate Immersion, U	Sea Coast U	Sea Coast D	Industrial U	Industrial D	Alternate Immersion U	Alternate Immersion D	Sea Coast D	Industrial U	Industrial D	2024 T351, U	7075 T651, U	7178 T651, U
1	Shot and Glass Bead Peening	3[a]	20	>35	>5	>5	4	>15	>15	>10	>10	1.6	4	16
2	Zinc Chromate Primer	4	5	11	>5	>5	10	8	5	2	4	1.7	—	4
3	Zinc Chromate Primer Plus Paint (Epoxy, Enamel, Polyurethane)	>140	>34	4	>5	2.5	40	10	5	>10	2.4	>35	45	>37
4	7072 Metal Spray	>140	>34	>34	>5	>5	>40	>16	>16	>10	>10	>35	30	>37
5	Paint (Epoxy, Polyurethane)	>150	12	3.5	>5	1.2	6	5	1.1	>10	1.0	>35	—	13
6	Conversion Coatings	—	—	—	—	—	0.14	—	—	>2	—	—	—	3
7	Anodized	0.2	—	—	—	—	—	—	—	—	—	—	—	—
	Hard	—	—	—	—	—	0.2	—	—	0.15	—	—	—	—
	Hot-water seal	0.2	—	—	—	—	—	—	—	—	—	—	—	—
	Dichromate seal	>13	—	—	—	—	6	—	—	1	—	—	—	25
8	Zinc Electroplate	>150	>34	>34	>5	>5	29	20	>20	>10	>10	>35	—	17
9	Aluminite 205	>150	6	12	1.3	1.2	0.7	0.5	0.4	0.5	0.6	60	—	19
	Aluminite 226 (modified)	0.5	2	1.6	0.5	0.7	14	2	1.2	1.1	1.3	0.4	—	0.3
10	Zinc Rich Paint	81	>34	8	>5	>5	>5	>15	>15	>10	6	27	—	8

Note: U = undamaged; D = damaged.

(a) Numbers indicate the degree of endurance relative to that of the unprotected alloy (expressed as 1).

TABLE II

Effect of shot peening on the short-transverse stress corrosion thresh-hold of Al-Zn-Mg alloy to DTD.5054 (Hawkes (8))

Preparation of specimen	UNPEENED		PEENED	
	Surface stress	SCC thresh-hold	Surface stress	SCC thresh-hold
Type A, Cut from fully heat-treated extrusion	-1½ hbar (-2 kips)	35 hbar (51 kips)	-19 hbar (-27 kips)	42 to 46 hbar (60 to 67 kips)
Type B, Cut and then heat-treated	-9 hbar (-13 kips)	23 to 26 hbar (32 to 34 kips)	-22 hbar (-31 kips)	34 to 42 hbar (49 to 60 kips)
Type C, Cut, bent, heat-treated and straightened.	+14 hbar (+20 kips)	20 to 26 hbar (29 to 38 kips)	-20 hbar (-29 kips)	31 to 35 hbar (45 to 51 kips)

(-ve values of surface stress are compressive, +ve tensile).

TABLE III

Results of stress-corrosion tests on short transverse Al-Cu-Mg specimens with 1/16 in. gaps in their sprayed coatings

Coating		Number of specimens	
Composition	Cu, %	Failed (> 10Xlife unprotected	Unfailed in 8 years (>100Xlife unprotected)
99.0% purity aluminium	0.02	1	1
99.5% purity aluminium	0.11 0.05 0.02 <0.01	8 5 4 0	4 7 10 2
99.97% purity aluminium	<0.01	0	2
Aluminium-1% zinc (based on 99.0% purity aluminium)	0.22	1	1
Aluminium-1% zinc (based on 99.5% purity aluminium)	0.12 0.05 0.02 <0.01	5 0 0 0	7 12 12 2
Pure zinc	0.02	0	4
Zinc-1% aluminium	<0.01	1	3

TABLE IV

Results of stress-corrosion tests on short transverse Al-Zn-Mg specimens with sprayed metal coatings

Coating		Number of specimens failed			Number of specimens unfailed in 8 years (> 250xlife unprotected)
Composition	Cu,%	<5xlife unprotected	5-10xlife unprotected	10xlife unprotected	
Complete					
99.5% purity aluminium	>0.05 0.02-0.05 <0.01	4 2 0	0 1 0	8 28 8	0 1 0
99.97% purity aluminium	<0.01	1	0	0	6
Aluminium-1% zinc	>0.05 0.02-0.05 <0.01	5 10 0	5 1 0	18 10 4	0 3 4
Pure zinc	0.02	2	1	7	10
Zinc-1% aluminium	<0.01	1	1	3	7
With 1/16" gaps					
99.5% purity aluminium	>0.05 0.02-0.05 <0.01	4 8 0	0 1 1	7 17 1	0 0 0
99.97% purity aluminium	<0.01	0	0	2	0
Aluminium-1% zinc	>0.05 0.02-0.05 <0.01	6 14 0	5 3 0	5 7 2	0 0 0
Pure zinc	0.02	3	0	1	0
Zinc-1% aluminium	<0.01	0	0	1	3

TABLE V

Results of Atmospheric exposure tests, U.K. Black tests (ref.12) Transverse specimens under four point bending

Material	Stress % 0.1% P.S.	LIVES - days		
		Untreated	Anodised*	An. & painted
DTD.363A, Al-Zn-Mg-Cu-Cr extrusion	90 70	33,108,261U 109,114,201U	34,49,116 68,123U,123U	75,167U,167U -
DTD.683, Al-Zn-Mg forging	84 55 25 15 8	45,69 110,144 543,557 1722,1760 3386U,3386U	44,45 87,103 320,382 1027,1081 3034,3386U	33,76 170,224 773,1033 2184,2814 3386U,3386U
DTD.683 forging	90 70 55 40	4,5,5 6,9,9 22,22,31 41,93,129	4,5,6 6,8,9 12,19,24 34,41,56	5,6,7 10,14,48 13,15,18 44,57,138
DTD.683 extrusion	90	234U,234U,234U	234U,234U,234U	-
DTD.683 extrusion	90	119,120,234U	44,203U,203U	-

* Chromic acid process

TABLE VI

Effect of surface mechanical treatments on stress corrosion life of 4340 steel (Davies, 22)

Short transverse bent beam specimens, alternate immersion in $3\frac{1}{2}$% sodium chloride solution

Surface preparation	Surface roughness	Surface compressive stress hbar	Time to failure, hours (Three specimens each)
Ground	23	36	13 to 40
Face milled	14	45	34 to 218
Chemically milled	300	31	$\frac{3}{4}$ to $3\frac{3}{4}$
Electropolished	95	9	$1\frac{1}{4}$ to $5\frac{1}{2}$
Grit blasted	126	63	40 to 48
Shot peened	24	59	443 to 845

1 hbar = 1450 lb/in^2.

TABLE VII

Stress corrosion tests on very strong steels

Total immersion in 3% NaCl aq.

Steel	Plain or notched	Applied stress	Life to failure, hours				
			Unprotected	Cd plated	Cd/Ni plated	Ph & paint *	Chromised
18 Ni maraging T.S. 175 hbar	Plain	98% 0.1% P.S.	15,38,214				
		95% "	735,864	6,22,1440, 2880		7300UB	85,92,140
		90% "	1040,1260, 1502	8,11,7200UB			
		75% "	48,2400UB	50,3000		6600UB	
	Notched	95% NTS				304	
		85% "	720UB, 950UB				
5Cr-Mo-V T.S. 231 hbar	Plain	97.5% 0.1% P.S			120,330		
		90% "	1, $1\frac{3}{4}$, 3	722,1128			
		75% "	$\frac{3}{4}$, 1, $1\frac{1}{4}$				
		50% "	$\frac{1}{2}$, $3\frac{1}{2}$, 7				
	Notched	65% NTS	$1\frac{1}{2}$				
		50% "	204,240				

* Blasted with alumina, phosphate treated by the Granodraw 5 process, chromated epoxy primer and epoxy finish.

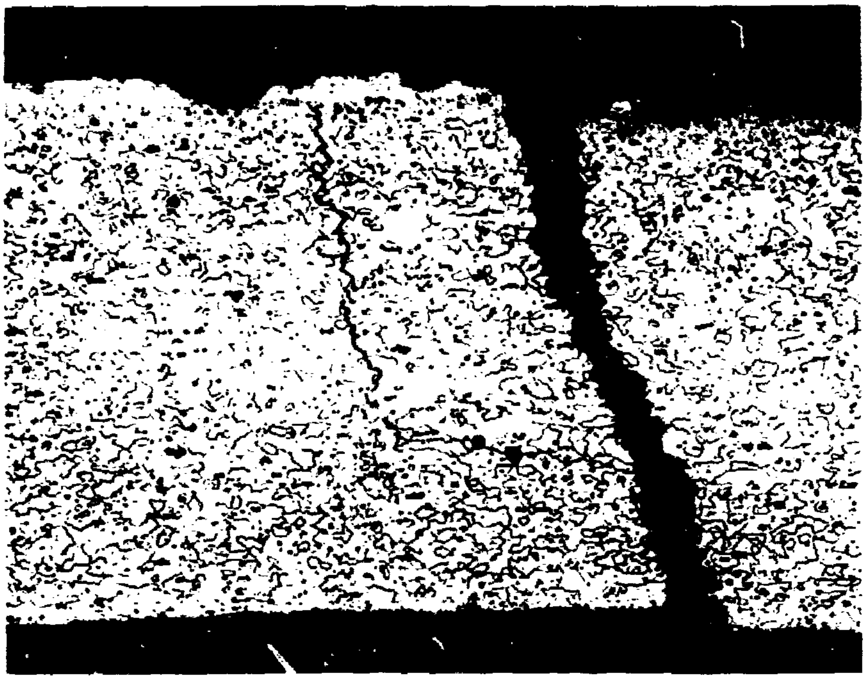

Fig. 1 Stress corrosion cracking in corroded clad Al-Zn-Mg alloy sheet

METHODS OF COMMUNICATION ABOUT PREVENTIVE ACTION

The Strategy of Stress Corrosion Cracking Prevention

by

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METHODS OF COMMUNICATION ABOUT PREVENTIVE ACTION

The strategy of stress corrosion cracking prevention

SUMMARY

Stress corrosion cracking cannot be prevented by only studying the theoretical backgrounds of it. A well organized system of communication is required in order to guarantee that all factors, that play a rôle in stress corrosion cracking, receive the required attention during the proper phase of the complete chain of activities from design through production to the operation of the aircraft.
The feed back of service experience is considered as vital for the succes of the action.
A diagram is given, showing the flow of information to the personnel concerned.

1. INTRODUCTION

The prevention of stress corrosion cracking in the structures of aerospace vehicles is a somewhat similar problem as the prevention of a disease that is dangerous and contagious for human beings. Some centuries ago the european population faced the danger of a multitude of such diseases. The fact that to-day the greater part of these have been banned out of our communities is not only the result of the thorough studies about the basic features of these diseases.
The fact that Robert Koch discovered the bacillus tuberculosis was not sufficient to fight tuberculosis. A large scale system of measures was required to achieve the present satisfactory situation. Systematic radiographic lung inspection of the complete population, advanced medical treatments of patients and even complete health control of the foodsupplying cattle, are only a few of the measures, required to keep this disease under control. Such preventive measures could not be effective unless they were supported by an efficient network of communication and information to the proper people. The know-how about keeping the cattle tuberculosis free was directed to the farmers and in a language that could be understood by farmers, etc..

The above quite elaborate example was used to explain that, in the similar case of stress corrosion cracking prevention, proper information should be supplied at the proper level.

In the following some suggestions are given how this could be effectively organized.

2. PROBLEM ANALYSIS

2.1. Stress corrosion parameters

The following factors are known to play a rôle in stress corrosion cracking:

(1) The type of alloy;

(2) The type of heattreatment;

(3) The type of semi-finished product : plate;
bar;
extrusion;
forging,

(4) The geometry of the semi-finished product, including the influences of fibre-directions in the end product.